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H. N. BOWEN

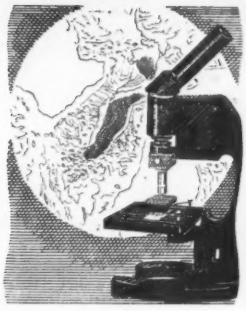
HUBERT NORMAN BOWEN, eldest son of William Bowen, Director and Secretary of the Copper Miners Tinplate Company Limited, was born in Cwmavon, Glamorgan, in 1904. He was educated at the Port Talbot County School and at the Technical College, Swansea, where he passed the final examination of the City and Guilds (London) Institute in Iron and Steel Manufacture.

Mr. Bowen took up his first appointment in 1920, when he joined the Laboratory Staff of the Port Talbot Steel Company. In 1927 he was loaned to the Copper Miners Tinplate Company Limited to establish and operate a laboratory at their works. On returning to the Port Talbot Steel Works he became Charge Controller at the No. 1 Melting Shop. In 1930 he was appointed Personal Clerk to the Chief Metallurgist of Guest Keen Baldwins Iron and Steel Company Limited. He became Assistant Sample Passer at the Margam Melting Shop of the Company in 1935, and after a temporary period as Sample Passer at their No. 2 Melting Shop at Port Talbot, returned to Margam as Sample Passer. He is now the Senior Sample Passer to the Steel Company of Wales Limited.

Mr. Bowen became a member of the Swansea Technical College Metallurgical Society (now the Swansea and District Metallurgical Society) in 1922. He was first appointed to the Council of the Society as a Students' Representative, and later as a full member, and was elected President for the session 1952-53. He was elected a Member of The Iron and Steel Institute in 1947.

H. N. Bowen
Honorary Member of Council, 1952-53

ANDREW CARNEGIE SCHOLARSHIP REPORT



Formation of Ferrite in Hypo-Eutectoid Plain Carbon Steels

By C. Margaretha Hickley, Dipl.ing., Ph.D., and J. H. Woodhead, B. Met.

SYNOPSIS

The rate of formation of ferrite during the isothermal decomposition of austenite has been studied in a series of hypo-eutectoid plain carbon steels. The progress of reaction has been followed by quantitative microscopy, and the separate effects of nucleation and growth have been estimated. 820

THE formation of ferrite in hypo-eutectoid steels has received little attention since Carpenter and Robertson[1] reported an extensive investigation using the 'gradient cooling' technique, although a number of isothermal transformation diagrams for hypo-eutectoid plain carbon steels have been published.[2–6] The effect of deoxidation practice on the rate of formation of ferrite was studied, within a limited temperature range, by McBride, Herty, and Mehl.[7] Lange[8] found that the maximum rate of isothermal formation of ferrite in steels containing 0·55 and 0·41% of carbon occurred at about 660° C., but his conclusions were based on a magnetic analysis which he admitted was not capable of great accuracy.

The mode of formation of ferrite has been reviewed by Mehl[9] and by Mehl and Dubé.[10] The effect of temperature, time, carbon content, and grain size on the morphology of ferrite was discussed, and a mathematical treatment of the thickening of a ferrite plate was presented. However, Mehl pointed out that "the study of rates of nucleation and growth in hypo-eutectoid steels is still in its infancy."

The aim of the present investigation has been to gain further insight into the transformation of austenite in hypo-eutectoid steels. Special attention has been paid to obtaining quantitative data on the formation of ferrite within the critical range, because in this range the times of transformation are reasonably long and the complication of pearlite formation is absent. Using these data, an attempt has been made to estimate the separate effects of nucleation and growth during the formation of ferrite.

EXPERIMENTAL PROCEDURE

General

Most of the work was carried out on four commercial hypo-eutectoid steels of low silicon and manganese contents, with the following analyses:

Steel	C, %	Mn, %	Si, %	P, %	S, %
C.1	0·08	0·31	0·18	0·013	0·013
C.2	0·19	0·37	0·31	0·024	0·007
C.3	0·33	0·32	0·14	0·017	0·013
C.5	0·47	0·35	0·10	0·021	0·008

For investigating the isothermal transformation of austenite the metallographic technique introduced by Davenport and Bain[11] was used. Specimens in the form of small discs (0·3 in. dia. × 0·07–0·11 in. thick) were austenitized in a vertical-tube furnace and rapidly transferred to a lead bath which was maintained at the required temperature within ± 2° C. After various times of treatment the specimens were quenched into water. The lead bath was frequently cleaned, and the specimens were agitated violently during the quench.

The austenitizing treatments used in the main series of experiments were: 4 min. at 1000° C. for steel *C*.1, 7 min. at 1000° C. for steel *C*.2, and 7 min. at 900° C. for steels *C*.3 and *C*.5. These treatments were chosen, on the evidence of preliminary experiments, to give about the same grain size (A.S.T.M. 5–6) in each case.

The relatively short austenitizing treatments avoided any great difficulty due to decarburization. It was found that the specimens could be treated in air at 1000° C. or lower without producing more than a very slight decarburization at the surface. For treatments at above 1000° C. an atmosphere of purified nitrogen was used.

All the transformed specimens were examined metallographically, the percentage of ferrite present being determined by point counting. The specimen

Manuscript first received 25th February, 1953, and in its final form on 27th July, 1953.

Dr. Hickley, who was at Sheffield University when the work described in the paper was carried out, is now with Imperial Chemical Industries Ltd. (Metals Division), Witton, Birmingham, and Mr. Woodhead is at Sheffield University.

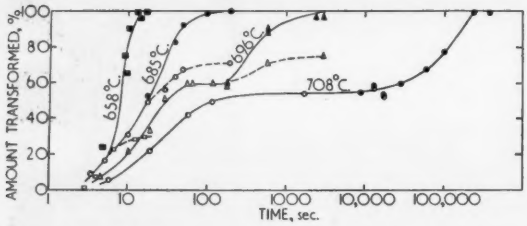


Fig. 1—Isothermal transformation of steel *C*.3 (open points refer to ferrite only, solid points to ferrite + pearlite)

preparation caused some tempering of the martensite, so that it was readily distinguishable from the ferrite. In each case 5–10 fields were studied, the number being chosen so as to give a standard error of not more than $\pm$ 1% of ferrite. The number of ferrite grains visible in the plane of polish after a grain-boundary etch was also counted.

Equilibrium Conditions within the Critical Range

Equilibrium conditions within the critical range can be approached either by the partial transformation of austenite to ferrite or by that of an aggregate of ferrite and cementite to austenite. Although the progress of the transformation of austenite was the major problem studied, it seemed desirable to investigate equilibrium conditions using both approaches. In consequence, specimens were given the standard austenitizing treatments, quenched into water, and then annealed for 5 hr. in a lead bath at the appropriate temperature. Finally, the specimens were water-quenched so that the austenite formed could be revealed as martensite. Specimens transformed directly from the austenitic condition were all given the same treatment of 5 hr. at the transformation temperature.

Iron–Carbon Alloys

Although commercial steels were used for the major portion of the investigation, it was considered desirable to obtain some results on reasonably pure iron–carbon alloys for comparison.

Alloys were prepared by gas-carburizing Swedish Armco iron which had been remelted under hydrogen. The method used was similar to that described by Digges.[12] The hydrocarbon used was ethyl benzene maintained at 18 $\pm$ ½° C. Purified hydrogen was bubbled through the ethyl benzene and became saturated with hydrocarbon vapour, corresponding to the vapour pressure obtaining at 18° C. The gas mixture then passed over 0·18-in. dia. bars of the iron in a vertical furnace maintained at 950 $\pm$ 5° C.

Care was taken to ensure homogeneity of the carburized bars. After an initial homogenizing treatment of 20 hr. at 1000° C., 0·2-g. samples were taken at 1·2-in. intervals along the bar. These samples were analysed for carbon by an absorptiometric colour method. The only pieces accepted were those having a carbon content within 0·02% of the mean value. Such pieces were then rehomogenized for 6 days at 1000° C. to eliminate residual cross-sectional inhomogeneity. All homogenizing treatments were performed with the specimens sealed in evacuated silica tubes. No oxidation or decarburization could be observed at the end of the treatments.

All the iron–carbon alloys contained about 0·05% of silicon, 0·15% of manganese, 0·020% of phosphorus, and 0·009% of sulphur. The carbon contents were as follows:

Material:	Basis Iron	*PA*.1	*PA*.2	*PA*.3	*PA*.4
Carbon, %:	0·03	0·10	0·17	0·28	0·37

Specimens of alloys *PA*. 1 and *PA*. 2 were austenitized for 7 min. at 1000° C., and specimens of *PA*. 3 and *PA*. 4 for 7 min. at 900° C. The resulting austenite grain sizes were similar to those produced in the commercial steels by the standard treatments (A.S.T.M. 5–6). In a few cases the progress of the transformation was studied, but most of the specimens were used for investigating the equilibrium amount of ferrite in the intercritical temperature range. As with the steels, the specimens were either transformed directly from the austenitic condition by holding for

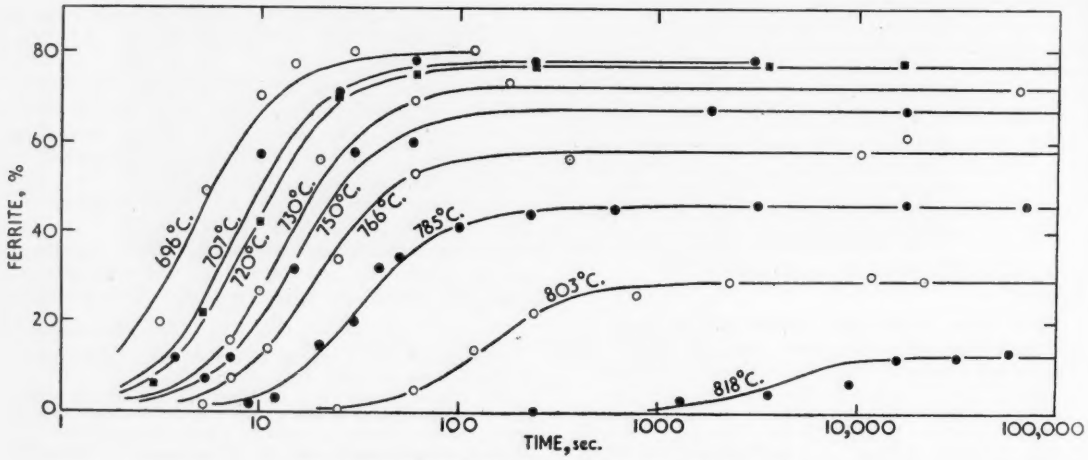


Fig. 2—Isothermal formation of ferrite in steel *C*.2

5 hr. in the lead bath or were first quenched into water and were then annealed in the lead bath for 5 hr. at the appropriate temperature.

RESULTS

Amounts of Ferrite

The characteristic behaviour of a hypo-eutectoid steel on transforming at temperatures below A_1 is shown in Fig. 1, where the fraction of austenite transformed for steel *C*.3 is plotted against time for a series of constant temperatures. At temperatures close to A_1 (696° and 708° C. in this case), the formation of ferrite reaches a standstill before any pearlite is formed. However, further free ferrite is formed during the pearlite reaction; this phenomenon is apparent at 696° C. At a slightly higher temperature (708° C.) this second increase in the amount of ferrite was readily observable, but it is not shown on the curve because a patchy distribution of the pearlite made quantitative estimation difficult. At lower temperatures (658° and 685°) the pearlite reaction starts earlier and overlaps the first stage in the formation of ferrite.

In Fig. 2 the formation of ferrite only is shown for steel *C*.2 at temperatures below the upper critical. The lowest temperatures included are those just below A_1, at which the formation of ferrite reaches a standstill before the pearlite reaction starts. The amounts of ferrite shown for these temperatures are those corresponding to the 'steady state' and do not include the increase that occurs during pearlite formation. All the points on this graph are experimentally determined, but the curves have been fitted using an equation which will be discussed later. Similar curves were obtained for steels *C*.1, *C*.3, and *C*.5, but Fig. 2 may be regarded as typical. The results showed that for all steels a decrease in transformation temperature leads to an increase in the rate of transformation and in the final amounts of ferrite formed.

From the isothermal reaction curves the steady-state amounts of ferrite have been evaluated for each steel and are plotted against temperature in Fig. 3. As the temperature is lowered, there is an increase in the steady-state amount of ferrite, and this increase continues below the A_1 temperature. The amounts of ferrite present in specimens transformed completely at constant temperatures below A_1 are also plotted in Fig. 3. As would be expected, the fully transformed specimens show a decrease in the amount of ferrite formed as the reaction temperature is lowered. The curves for the steady-state and fully transformed conditions for any particular steel do not meet but overlap. This overlapping is not in agreement with the smooth relationship between amount of ferrite and temperature postulated by Mehl.[9] The difference between corresponding curves is a measure of the amount of free ferrite formed during the pearlite reaction.

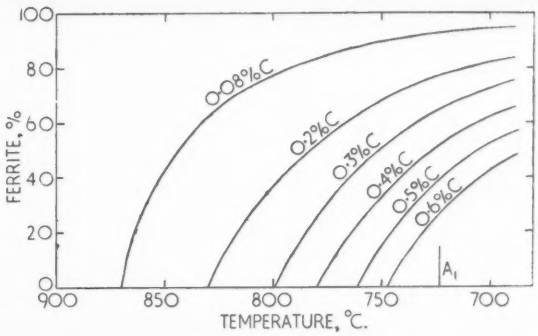


Fig. 4—Equilibrium amounts of ferrite as calculated from the iron–carbon diagram

For comparison, the final amounts of ferrite to be expected in the absence of pearlite have been computed from the iron–carbon equilibrium diagram (with extended phase boundaries) and are shown in Fig. 4. The relevant part of the iron–carbon diagram used for these computations is shown in Fig. 5. The data were taken from various sources.[13–15]

Comparison of Figs. 3 and 4 shows good agreement between theoretical and experimental results for the two low-carbon steels, but for the two higher-carbon steels the amount of ferrite formed is appreciably less than equilibrium considerations suggest.

Further investigation of this matter gave the results shown in Fig. 6. These curves show the amounts of

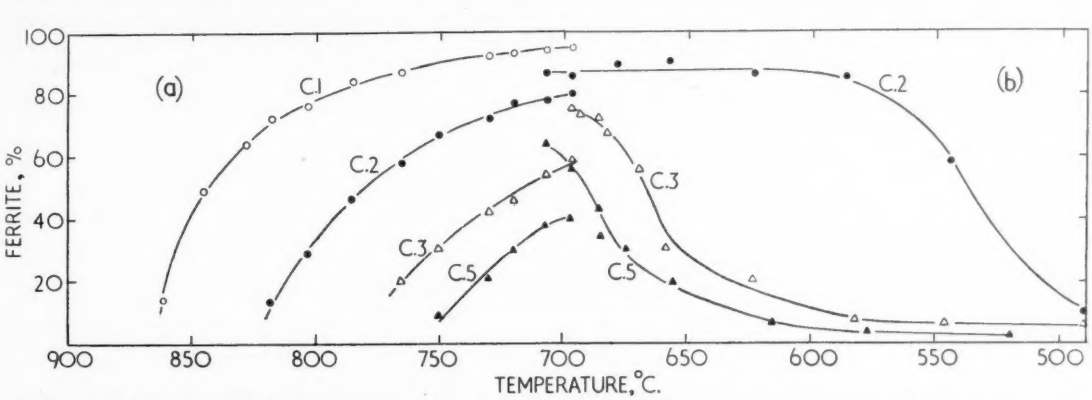


Fig. 3—Percentage of ferrite formed as a function of transformation temperature: (*a*) Steady-state amounts, (*b*) final amounts in fully transformed specimens

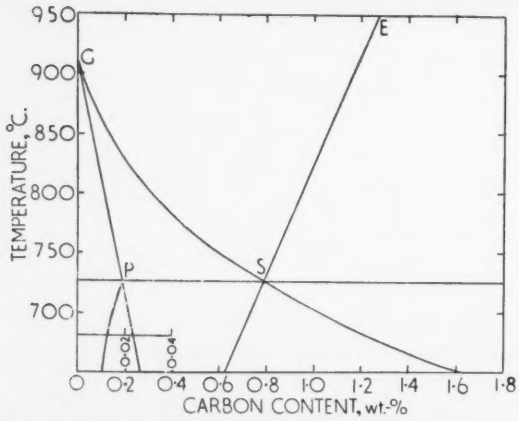


Fig. 5—Iron–carbon equilibrium diagram with extended phase boundaries. The scale for the solubility of carbon in ferrite is ten times that for the other phase boundaries

ferrite resulting from the 5-hr. anneal at various intercritical temperatures (*a*) after water-quenching from the standard austenitizing temperature, (*b*) directly after austenitizing at 900° C., and (*c*) after austenitizing at 1000° C.

These experiments show that the final amount of ferrite formed above A_1 is to some extent structure-sensitive. There is appreciably more ferrite formed in the quenched and reheated specimens than there is in those transformed directly from austenite. Also, the lower austenitizing temperature gives rise to more ferrite than does the higher temperature, but the difference is slight except in the case of steel *C*.2. The 5-hr. direct-transformation treatments give amounts of ferrite insignificantly different from the steady-state amounts plotted in Fig. 3.

The results of experiments carried out on the series of pure iron–carbon alloys are shown in Fig. 7. The difference between the amounts of ferrite formed directly from austenite and from martensite is small; only one curve is drawn for each alloy. Agreement

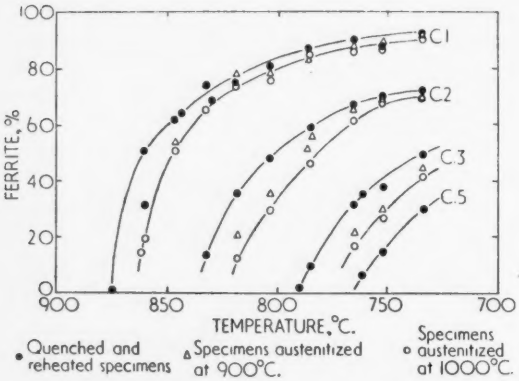


Fig. 6—Amounts of ferrite in the hypo-eutectoid steels after a 5-hr. treatment as a function of transformation temperature

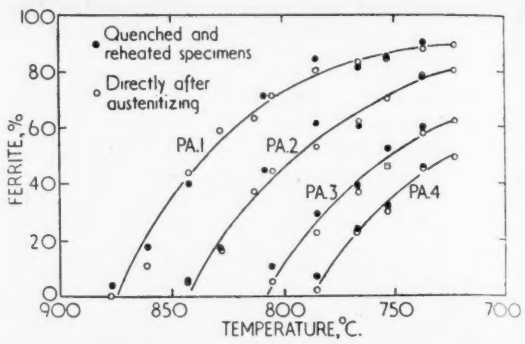


Fig. 7—Amounts of ferrite in the iron–carbon alloys after a 5-hr. treatment as a function of transformation temperature

with the theoretical curves is good for the two lower-carbon alloys. For the higher-carbon alloys the agreement with the theoretical values is fair and it is much better than for the commercial steels; however, at lower temperatures there is less ferrite than would be expected from equilibrium conditions.

Size of Ferrite Particles

As mentioned previously, not only was the amount of ferrite determined micrographically, but also the number of ferrite grains present in the plane of polish was counted. From these results the radius $\bar{\rho}$ of an average-sized patch (assumed to be produced by a spherical particle) was computed. It was found that $\bar{\rho}$ plotted against log time gives an apparent straight-line relationship until the increase in $\bar{\rho}$ ceases. Therefore,

$$\frac{d\bar{\rho}}{d(\log t)} = C,$$

$$\frac{d\bar{\rho}}{dt} = \frac{C}{2{\cdot}3}\cdot\frac{1}{t} \qquad \text{.....(1)}$$

where C is a constant and t = time. Thus $d\bar{\rho}/dt$, the rate of increase in radius of an average patch, is inversely proportional to time (for constant temperature). To compare the effects of variation in temperature and carbon content on $d\bar{\rho}/dt$, this function was determined at a specific value of $\bar{\rho}$, namely, 4×10^{-4} cm. This value was chosen because it was near or within the experimentally found range of patch sizes, for all steels at the various

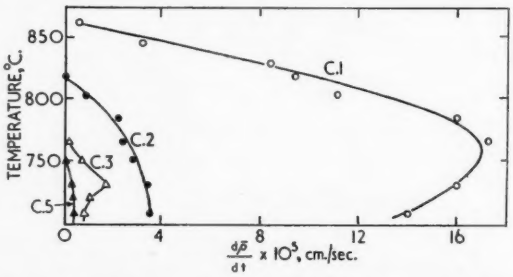


Fig. 8—Rate of increase in radius of an average-sized patch $d\bar{\rho}/dt$ when $\bar{\rho} = 4 \times 10^{-4}$ cm. as a function of transformation temperature

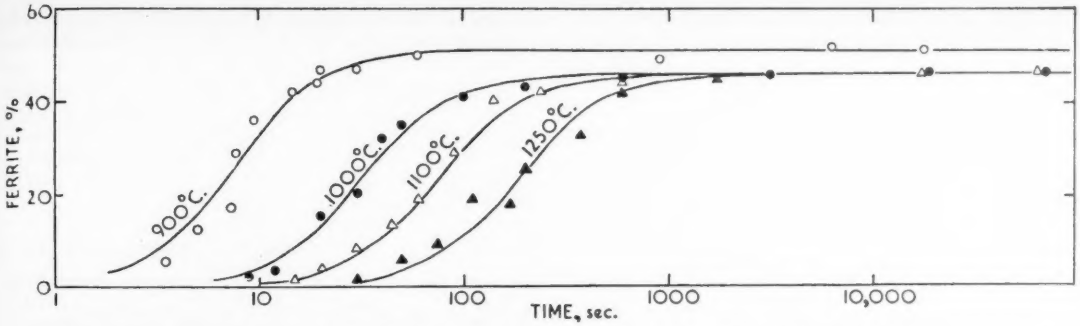


Fig. 9—Isothermal formation of ferrite in steel *C*.2 transformed at 785° C. after various austenitizing treatments

temperatures, at a comparatively early stage of transformation, *i.e.*, before impingement of neighbouring patches becomes appreciable.

The results are shown in Fig. 8, which gives a qualitative picture of the rate of growth. This increases as carbon content decreases and as temperature falls, possibly passing through a maximum with falling temperature. The shape of the curve for steel *C*.3 is probably a result of experimental error. No quantitative conclusions about growth rates can be drawn from these curves, because $d\bar{p}/dt$ is affected by the rate of nucleation and by the extent to which transformation has occurred as well as by the rate of growth.

Effect of Varying Austenitizing Treatment on Transformation

Although most of the experiments were carried out using austenitizing temperatures giving approximately the same austenite grain sizes for all the steels, it seemed desirable to obtain some information about the effect of austenitizing temperature on the progress of the transformation. A series of experiments was therefore performed using the carbon steel *C*.2. Samples were transformed at one intercritical temperature (785° C.) after different austenitizing treatments. The fraction transformed is shown as a function of time in Fig. 9, and the variation of average particle radius with time is shown in Fig. 10. These results are discussed on p. 137.

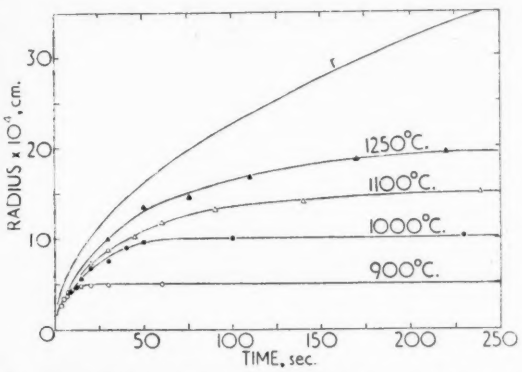


Fig. 10—Radius of an average-sized particle as a function of time for steel *C*.2 transformed at 785° C. after various austenitizing treatments. The calculated radius *r* of a single spherical particle is also shown

DESCRIPTION OF THE TRANSFORMATION IN TERMS OF NUCLEATION AND GROWTH

A complete account of the formation of ferrite in hypo-eutectoid steels requires a description of the shape, size, and number of ferrite particles existing at any instant during the transformation. In addition, the distribution of these particles relative to the pre-existing austenite structure should be described and any consistent orientation relationships should be specified. In the present work, attention has been concentrated upon the possibility of calculating nucleation and growth rates; consequently, the determination of the size and number of ferrite particles has received most attention. However, the shape and distribution of the particles could not be ignored, because measurements on the structure as observed in the plane of polish do not permit accurate quantitative evaluation of the size and number of particles in three dimensions, except for the case in which the particles are spherical.

Shape of Ferrite Particles

At high temperatures, near the respective A_3 points, ferrite forms as irregular polygonal grains (' blocky ferrite '), which lie on the austenite grain boundaries (*see* Fig. 13). At a low temperature (*see* Fig. 14) ferrite forms a complete envelope around the austenite grains, with plates growing towards the centre of a grain. The spacing and thickness of the ferrite plates decreases with temperature. There is a gradual change from the first type to the second with decrease in the temperature of transformation.

An intermediate case is represented by Figs. 15*a* and *b*, which show early and late stages in the transformation of the carbon steel *C*.2 transformed at 765° C. The angular shapes of the early precipitates become more rounded before the end of the reaction.

The shape of the ferrite particles is also affected by the austenitizing temperature; an increase in this temperature increases the tendency towards the formation of platelike particles.

Correlation between Structure in the Plane of Polish and in Space

The actual size and number of particles can be calculated from observations on a plane of polish

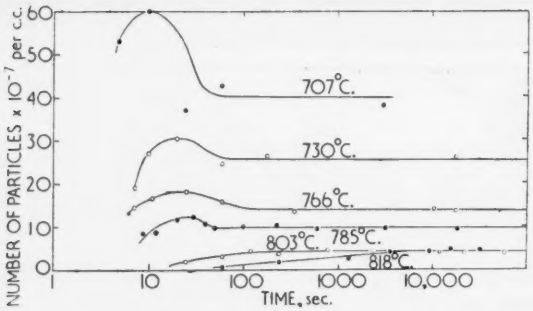


Fig. 11—Estimated number of particles as a function of time for steel *C.2*

when the particles are approximately spherical. For randomly distributed uniform-size spheres, Scheil[16] gave the expression

$$N_r = 2rM_r \quad \ldots\ldots(2)$$

where N_r = number of particles per unit area in the plane of polish, M_r = number of spheres per unit volume, and r = radius of a sphere. Using equation (2) as a basis, Scheil has developed a method of determining the size distribution and total number of particles in space from a knowledge of the distribution of patch radii.

For uniform-size spheres, the relationship between r and $\bar{\rho}$ (the radius of a patch of average area) can be obtained by equating the fraction of volume occupied by the spheres to the fraction of area occupied by the patches:

$$M_r \cdot \frac{4}{3}\pi r^3 = N_r \cdot \pi \bar{\rho}^2,$$

whence

$$r = \sqrt{\left(\frac{3}{2}\right)}\bar{\rho} \quad \ldots\ldots(3)$$

Introducing this value for r in equation (2),

$$M_r = N_r/2 \cdot 46\bar{\rho} \quad \ldots\ldots(4)$$

Number of Particles as a Function of Time and Temperature of Transformation

In the present work, $\bar{\rho}$ and the number of patches were obtained experimentally. Equation (4) was used to estimate the number of particles present after different times at each constant temperature. The distribution of patch radii was not determined because, as the particles at best only approximate to spheres, no real improvement in accuracy would have been obtained and the greatly increased work involved in such a determination would not be justified.

Typical results are plotted in Fig. 11. It is apparent that only the curves for the two highest transformation temperatures have the expected shape. This may be due to the fact that the particles are not spherical; at the lower transformation temperatures their shapes cannot be considered even to approximate to spheres. Qualitatively, Fig. 11 shows that the rate of nucleation must decrease rapidly with time.

A certain decrease in the number of particles with time could be caused by agglomeration of existing ferrite grains, but this would seem likely to be a time-consuming process.

As the transformation proceeds, the ferrite particles become more rounded and, at the end of an isothermal transformation, equation (4) should give a good approximation to the final number of particles. Such data are plotted in Fig. 12 for the different steels. Although these curves show the trend of the rate of nucleation with temperature, the final number of particles is also influenced by the rate of growth, which may vary appreciably.

Simple Reaction Equation for the Formation of Ferrite

The method that has been used most for the determination of rates of nucleation and growth, particularly for the austenite to pearlite transformation,[17] consists in measuring experimentally the radial rate of growth G and the time required for 50% transformation, and then calculating the rate of nucleation N from a reaction equation

$$f(t) = 1 - \exp(-\tfrac{\pi}{3} \cdot NG^3t^4) \quad \ldots\ldots(5)$$

where $f(t)$ is the fraction transformed and t is time.

The derivation[18] of this equation involved the following assumptions: (*a*) N and G remain constant during transformation; (*b*) nucleation is random; and (*c*) the reaction product forms true spheres up to the moment of impingement. Except for the constancy of G, which was experimentally verified, none of these assumptions is strictly valid for pearlite, but the deviations are small enough to justify the use of the method for obtaining a good approximation. More complex expressions have been derived for the case of grain-boundary nucleation,[18] but equation (5) has generally been preferred.

The general form of equation (5) is

$$f(t) = 1 - \exp(-kt^n) \quad \ldots\ldots(6)$$

where k and n are constant for any particular case. This expression may be derived as follows. It is first assumed that $f(t)$ is proportional to some power of time before impingement between growing particles takes place:

$$f(t) = kt^n,$$

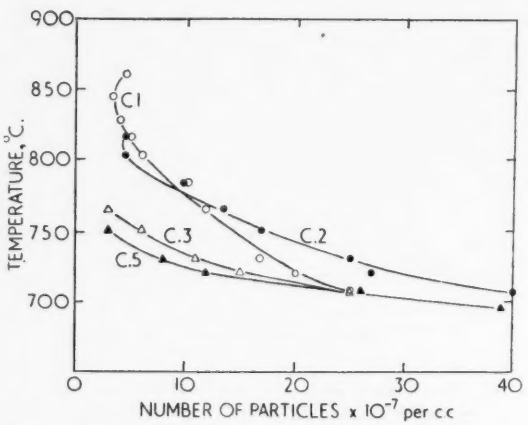
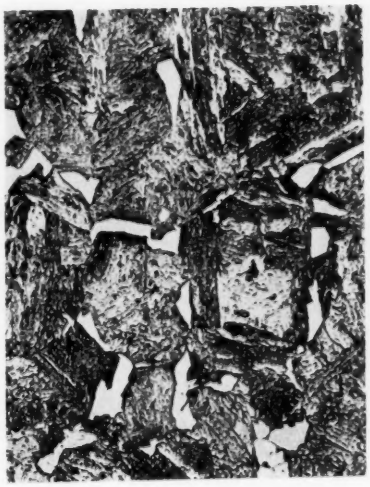


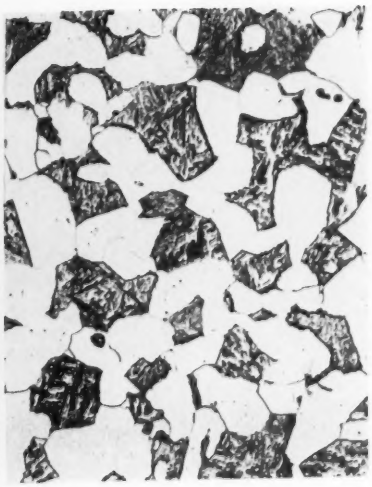
Fig. 12—Final number of particles as a function of transformation temperature

Fig. 13—Blocky ferrite in steel *C*.1 transformed at 845° C. for 120 sec. × 400

Fig. 14—Plate-like ferrite in steel *C*.3 transformed at 658° C. for 5 sec. × 400

(a)





(b)

Fig. 15—Steel *C*.2 transformed at 766° C. for (*a*) 7 sec. and (*b*) 5 hr. × 400

Fig. 16—Darkening of martensite adjacent to growing ferrite particles (steel *C*.2 transformed at 707° C. for 10 sec.) × 1000

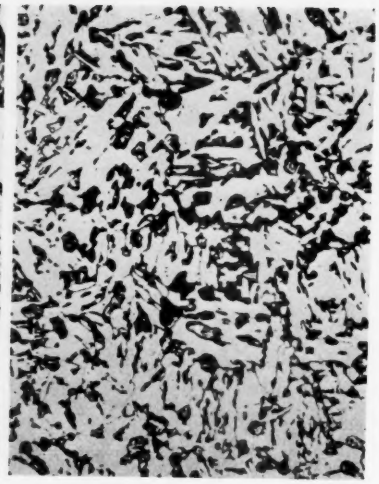


Fig. 17—Steel *C*.2 quenched and then annealed at 766° C. for 5 hr. × 400

Fig. 18—Pearlite and carbide in steel *C*.3 transformed at 708° C. Etched in alkaline sodium picrate solution × 400

and therefore the rate of transformation

$$\frac{df(t)}{dt} = nkt^{n-1}.$$

If, following the work of Johnson and Mehl,[18] impingement is taken into account by assuming that the rate of transformation is proportional to the fraction untransformed, $u(t)$,

$$\frac{df(t)}{dt} = nkt^{n-1}.u(t);$$

but $$\frac{df(t)}{dt} = \frac{-du(t)}{dt},$$

therefore $$\frac{du(t)}{u(t)} = -nkt^{n-1}dt.$$

By integration, and since $u(t) = 1$ when $t = 0$,

$$u(t) = \exp(-kt^n)$$
$$f(t) = 1 - \exp(-kt^n).$$

If this expression holds good, a plot of $\log \ln\left[\frac{1}{1-f(t)}\right]$ against $\log t$ should give a straight line.

Experimental results for the formation of ferrite from austenite were plotted on this basis, and examples are given in Fig. 19; $f(t)$ was taken as the fraction of ferrite relative to the final amount formed and not that relative to the total volume. It is apparent from Fig. 19 that at the later stages of transformation ($f(t)$ greater than about 0·7) equation (6) does not hold. This misfit is not surprising, because the impingement factor of equation (6) assumed random distribution of particles. In the present case the ferrite particles form in clusters around the austenite grain boundaries, and so the effect of impingement will be much greater.

To allow for this greater effect of impingement, assume arbitrarily that the rate of transformation becomes proportional to the square of the fraction untransformed. By analogy with the previous derivation,

$$\frac{df(t)}{dt} = nkt^{n-1}u(t)^2$$

or $$\frac{du(t)}{u(t)^2} = -nkt^{n-1}.dt,$$

whence, by integration,

$$\frac{1}{u(t)} = kt^n + C.$$

As before, $u(t) = 1$ when $t = 0$, and therefore $C = 1$. Thus

$$\frac{f(t)}{1-f(t)} = kt^n \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots (7)$$

If this expression holds, a plot of $\log \frac{f(t)}{1-f(t)}$ against $\log t$ should give a straight line. Such plots for some of the data are shown in Fig. 19. The equation fits the results to a higher percentage transformation ($f(t) \simeq 0{\cdot}9$) than does equation (6).

Accordingly, all the experimental results for the formation of ferrite were plotted as $\log \frac{f(t)}{1-f(t)}$ *versus* $\log t$. The values of n obtained (from the slope of the lines) varied between 1·6 and 3·0, but they were close to 2 in a great number of cases. The higher values were usually observed for fast reactions; the shorter the immersion time, the greater is the proportion of this which is taken up in reaching the transformation temperature. A correction employed would therefore lower the value of n for fast reactions, but it would have very little influence on slower

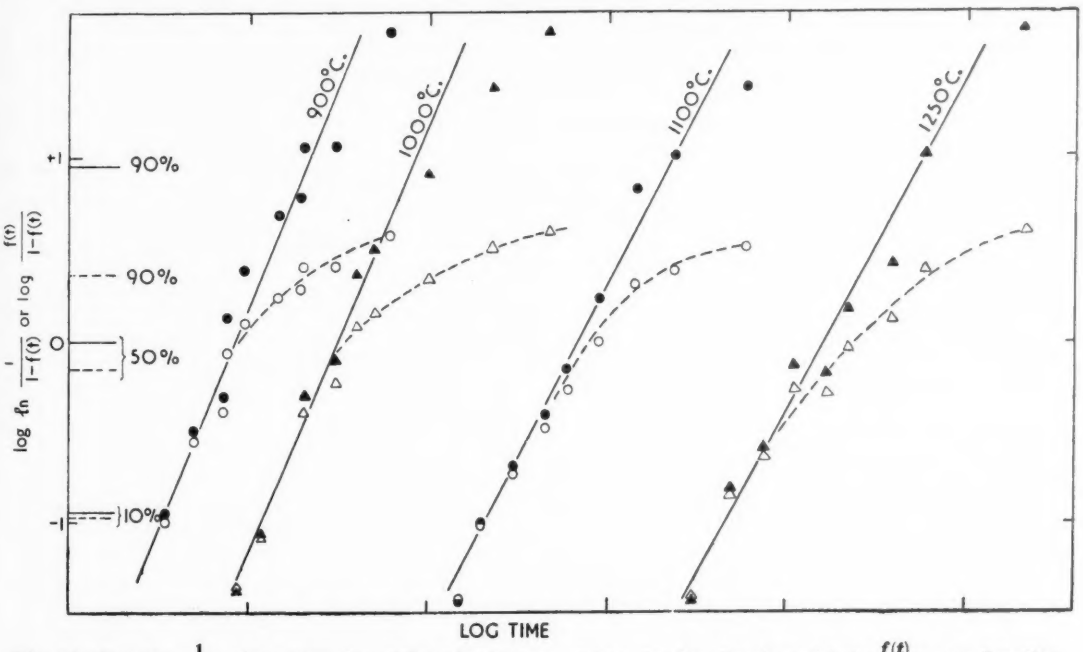


Fig. 19—Log ln $\frac{1}{1-f(t)}$ as a function of log (time) (open points, broken lines); and log $\frac{f(t)}{1-f(t)}$ as a function of log (time) (solid points, full lines). Results are for steel *C.2* transformed at 785° C. after various austenitizing treatments

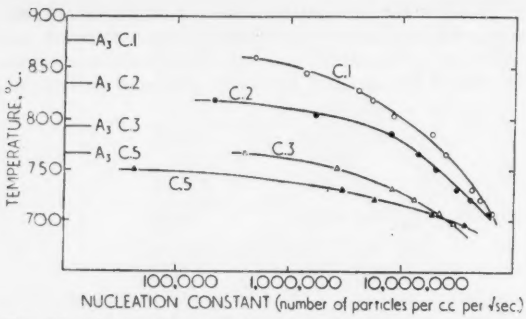


Fig. 20—Nucleation constants as a function of transformation temperature

transformations. In general, the formation of ferrite has been expressed by the equation

$$\frac{f(t)}{1-f(t)} = kt^2 \quad \ldots\ldots\ldots\ldots\ldots\ldots (8)$$

and the curves in Figs. 2 and 9 have been fitted to the experimental points using this expression. The values of k were obtained from the times corresponding to $f(t) = 0{\cdot}5$, when $\log k = -2 \log t$.

Effect of Nucleation and Growth on Transformation

An attempt was made to explain equation (8) in terms of the rate of nucleation and the rate of growth.

Zener[19] has derived an expression for the rate of growth of a spherical precipitate in a solid solution. This can be expressed as

$$r = gt^{\frac{1}{2}} \quad \ldots\ldots\ldots\ldots\ldots\ldots (9)$$

where r = radius of sphere, t = time, and g = a growth constant. Thus the volume of a sphere is proportional to $(t-t')^{3/2}$, where t' = time when the nucleus was formed. For a thin disc the volume is proportional to $(t-t')^{5/2}$, and for a rod the volume is proportional to $(t-t')^{4/2}$.[20]

Although the ferrite particles are not spherical, in the temperature range between A_3 and about 700° C. the shape approximates more closely to a sphere than to a disc or a rod. Accordingly, assuming spherical growth, for a particle formed at time T the radius is

$$r = g(t-T)^{\frac{1}{2}}$$

and the volume is

$$v = \frac{4}{3}\pi g^3 (t-T)^{3/2}.$$

The rate of increase in volume is

$$\frac{dv}{dt} = 2\pi g^3 (t-T)^{\frac{1}{2}}.$$

Although the present experiments permitted no quantitative determinations of the rate of nucleation, this was qualitatively shown to decrease with time. However, to account for the exponent of 2 in equation (8), it has been assumed that the rate of nucleation, before impingement, is inversely proportional to the square root of time. Then the number of particles formed between T and $(T+dT)$ is $(c/T^{\frac{1}{2}})dT$, where c is a constant. The rate of increase in volume of ferrite particles formed between T and $(T+dT)$ is then

$$2\pi g^3 c \frac{(t-T)^{\frac{1}{2}}}{T^{\frac{1}{2}}} dT,$$

and for all the particles formed up to time t

$$\int_0^t 2\pi g^3 c \frac{(t-T)^{\frac{1}{2}}}{T^{\frac{1}{2}}} dT = \pi^2 g^3 ct.$$

The effective rate of increase in volume divided by the fraction of total ferrite formed (a) is equal to the rate of transformation. The increase in volume is influenced by the impingement of neighbouring particles; if this is corrected for in the same way as in equation (7),

$$\frac{df(t)}{dt} = \frac{\pi^2 g^3 c}{a} \cdot t[u(t)]^2$$

and similarly

$$\frac{f(t)}{1-f(t)} = \frac{\pi^2 g^3 c}{2a} \cdot t^2 \quad \ldots\ldots\ldots\ldots\ldots\ldots (10)$$

Comparison with equation (8) gives

$$k = \frac{\pi^2 g^3 c}{2a} \quad \ldots\ldots\ldots\ldots\ldots\ldots (11)$$

In the case of randomly distributed particles, if it is assumed that the rate of nucleation (per unit volume untransformed matrix) is constant with time, the total number of particles formed during the reaction is given by the expression

$$\int_0^\infty Nu(T)dT.$$

In the present case it was assumed that the rate of nucleation $= c/\sqrt{\text{time}}$, in regions unaffected by impingement. By analogy with the above, the total number of particles formed (Z) is

$$Z = \int_0^\infty \frac{c}{T^{\frac{1}{2}}} u(T)dT.$$

Since $u(T) = 1/(1+kT^2)$, from equation (8),

$$Z = \int_0^\infty \frac{c}{T^{\frac{1}{2}}} \cdot \frac{1}{1+kT^2} dT.$$

Integrating,

$$Z = \frac{c\pi}{k^{\frac{1}{4}}\sqrt{2}} \quad \ldots\ldots\ldots\ldots\ldots\ldots (12)$$

If the same impingement factor is used as in deriving equation (7),

$$Z = \int_0^\infty \frac{c}{T^{\frac{1}{2}}} [u(T)]^2 dT = \frac{3}{4} \cdot \frac{c\pi}{k^{\frac{1}{4}}\sqrt{2}},$$

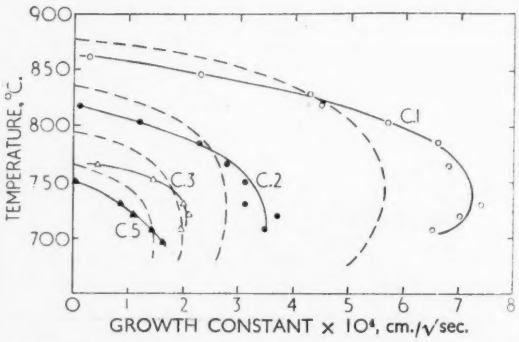


Fig. 21—Radial growth constants as a function of transformation temperature. The full lines are based on experimental results, the broken lines have been calculated theoretically

which differs from equation (12) only by the factor $\frac{3}{4}$.

The fraction transformed as a function of time and also the final number of particles have been determined experimentally. Thus k and Z are known and the nucleation constant c can be calculated using equation (12). Knowing c and k, the value of the radial growth constant g can then be obtained using equation (11).

The values of the nucleation and growth constants thus calculated for the steels investigated are shown in Figs. 20 and 21 as functions of temperature.

Assessment of Method Used for Calculation of Rates of Nucleation and Growth

The derivation of equations (11) and (12) involves the following assumptions: (*a*) The ferrite particles grow as spheres, the equation $r = gt^{\frac{1}{2}}$ being valid; (*b*) the rate of nucleation is proportional to $1/\sqrt{\text{time}}$; and (*c*) the rate of transformation is proportional to the square of the fraction untransformed.

The series of experiments carried out with the 0·2% C steel (*C*.2) at 785° C. can serve as a test of the method of calculating the rates of nucleation and growth suggested above. Table I gives the details of the experiment and of the calculated values of the nucleation and growth constants. The value of g should be the same in each case. The values vary by only $\pm$ 7%, which, in view of the assumptions involved, is a satisfactory agreement.

DISCUSSION

Reaction Equation for Ferrite

Examination of the curves for ferrite formation in Figs. 2 and 9 shows that the agreement between experimental points and fitted curves varies. In many cases a better fit could have been obtained by adjusting the value for n in equation (7), but, as the calculations of the rates of nucleation and growth have been based on the assumption that $n = 2$, this value has been used throughout.

The contribution of growth to n works out as 3/2 because spherical growth was assumed, and the contribution of nucleation was taken as 1/2. With decrease in temperature the particles become less spherical, and it would therefore be more correct to assume the effect of growth on n to be somewhat larger than 3/2 and that of nucleation somewhat smaller than 1/2. However, the mathematical treatment would then become complicated; the present simplified method should serve for purposes of comparison.

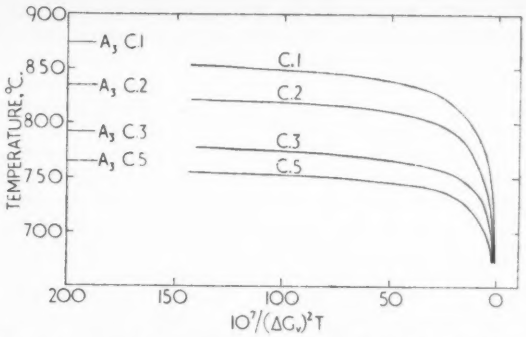


Fig. 22—Effect of temperature on the equilibrium number of nuclei as shown by $1/[(\Delta G_v)^2 T]$

All the transformation curves have been calculated assuming a maximum rate of nucleation at zero time and an initial maximum rate of growth, both these decreasing with time. The incubation periods which can be observed, particularly at higher temperatures, merely indicate that the rates of nucleation and growth are slow and that time is required before an appreciable amount of ferrite is formed.

Rates of Nucleation

In all the cases investigated, nucleation was found to occur preferentially at the austenite grain boundaries, the increase in number of visible particles with time being relatively small. Presumably the most probable sites of nucleation at the grain boundaries are being rapidly used up.

With an increasing amount of undercooling there is an increase in the rate of nucleation, as shown by the variation of the constant c with temperature (*cf.* Fig. 20). The increase is less rapid at lower temperatures, particularly for the two low-carbon steels. This is in accordance with theoretical expectations for homogeneous nucleation (*see* Fig. 21, which gives a picture of the effect of increased degree of undercooling on the equilibrium number of nuclei).

The equilibrium number of nuclei N_n is proportional to $\exp(-G_c/kT)$, where G_c is the energy necessary to form the particle of critical size and k is Boltzmann's constant. If the interfacial energy per unit surface area is considered to be independent of temperature, G_c is proportional to $1/(\Delta G_v)^2$, where ΔG_v is the energy liberated per unit volume of precipitate.[14] Thus $\log N_n$ is a function of $1/[(\Delta G_v)^2 kT]$.

Table I

STEEL *C*.2 TRANSFORMED AT 785° C.

Austenitizing Temperature, °C.	A.S.T.M. Grain Size	n	k*	Final No. of Particles per c.c.	Nucleation Constant c, nuclei/c.c./$\sqrt{}$ sec.	Growth Constant g, cm./$\sqrt{}$ sec.
900	8	2·4	$1{\cdot}8 \times 10^{-2}$	8×10^{8}	$1{\cdot}3 \times 10^{8}$	$2{\cdot}4 \times 10^{-4}$
1000	5	2·3	$1{\cdot}1 \times 10^{-3}$	1×10^{8}	$8{\cdot}2 \times 10^{6}$	$2{\cdot}3 \times 10^{-4}$
1100	$4\frac{1}{2}$	2·0	$1{\cdot}8 \times 10^{-4}$	3×10^{7}	$1{\cdot}6 \times 10^{6}$	$2{\cdot}2 \times 10^{-4}$
1250	2	1·8	$3{\cdot}1 \times 10^{-5}$	9×10^{6}	$3{\cdot}0 \times 10^{5}$	$2{\cdot}1 \times 10^{-4}$

* Calculated assuming $n = 2$

Figure 22 shows a plot of $10^7/[(\Delta G_v)^2 T]$ against temperature for iron–carbon alloys of the same carbon contents as the steels investigated. The values of ΔG_v for the initial formation of ferrite from supersaturated austenite have been calculated using data from various sources.[21–23]

The lower the carbon content the greater is the nucleation constant at a certain temperature. A more rigorous comparison of the different carbon steels would require the same state of homogeneity and grain size of the austenite in each case. These conditions might not obtain experimentally; a small variation in homogeneity can have a noticeable effect on nucleation, as is the case for pearlite.

Rates of Growth

The radial growth constants as computed from the experimental data have been compared in Fig. 21 with values calculated using an expression given by Zener[19]:

$$g = D^{\frac{1}{2}}.K_3(n_\infty - n_1)^{\frac{1}{2}}/(n_0 - n_\infty)^{\frac{1}{2}} \quad \ldots\ldots\ldots(13)$$

where n_1 = concentration of the solute in the parent phase at the interface
n_∞ = concentration of the solute in the parent phase far from the interface
n_0 = concentration of the precipitate
D = diffusion coefficient, assumed independent of concentration
K_3 = a constant which varies from 1·4 to 2·4 as n_∞ varies from n_1 to n_0.

It has been assumed that the iron–carbon equilibrium diagram is applicable to the steels investigated, and that the equilibrium concentrations are maintained at the interfaces. Then

n_1 = concentration given by the A_3 line or its extension
n_0 = carbon content of the ferrite
n_∞ = carbon content of the steel.

The diffusion coefficients used were calculated from data[24] for a concentration of 1 at.-% of carbon for steels *C*.1 and *C*.2, 1·5 at.-% of carbon for steel *C*.3, and 2 at.-% of carbon for steel *C*.5. Although the diffusion coefficient is assumed to be constant in Zener's expression, this is, of course, an approximation, and it seemed best to use the relevant value of *D* for each steel.

That the austenite is enriched in carbon in the vicinity of its interface with ferrite is shown clearly in Fig. 16. After partial transformation the specimen was quenched so that the residual austenite transformed to martensite. This martensite etches more darkly where it is of higher carbon content.

The experimental curves for the growth constant show a slower rate of growth at higher temperatures and a faster rate of growth at lower temperatures than do the theoretical curves. The discrepancy at lower temperatures is possibly due to the tendency of the ferrite particles to assume more platelike shape, in which case the increase in volume with time would be greater than for spherical growth.

The slower rate of growth found at higher temperatures could be due to the fact that the steels investigated contained small amounts of alloying elements. The amounts of alloying elements commonly found in plain carbon steels are thought to have little effect on the diffusion coefficient for carbon in austenite, but their influence on the equilibrium concentrations of carbon in austenite and ferrite are not accurately known. Minor contents of alloying elements may slow down the growth rate of ferrite by retarding the actual γ–α interface reaction.[9] Also, the diffusion of an alloying element during the formation of ferrite could decrease the rate of transport of carbon by altering the activities of carbon in the surrounding austenite.

On the whole, the agreement is as good as could be expected in view of the approximations introduced.

Final Amounts of Ferrite

The increase in the steady-state percentage of ferrite with decrease in temperature continues below A_1, as would be expected from a consideration of the extended phase boundaries. Although the agreement with the theoretical curves (*cf.* Figs. 3 and 4) is not so good for steels *C*.3 and *C*.5, the increase in amount of ferrite below A_1 is apparent and seems to provide evidence for the validity of the assumption of extended phase boundaries.

The poor agreement between experiment and theory for steels *C*.3 and *C*.5 is not fully understood. It could be caused by a very slow rate of approach to equilibrium or because the actual equilibrium conditions differ from those assumed. The small amounts of alloying elements present are not expected to alter the equilibrium conditions to a sufficient extent to account for the discrepancy. Confirmation of this supposition was obtained through the good agreement between the A_3 temperatures as estimated from the iron–carbon equilibrium diagram and those derived from Fig. 6. The latter were obtained by extrapolating to 0% ferrite the experimental curves relating the amount of ferrite to the temperature at which a quenched specimen was reheated for 5 hr.

The agreement between experiment and theory for the pure alloys *PA*.3 and *PA*.4 (*see* Fig. 7) is better than for steels *C*.3 and *C*.5, but even for these there is at lower temperatures less ferrite than would be expected from equilibrium considerations.

Appreciably more ferrite was found in specimens reheated from a martensitic structure than in those transformed directly from austenite. In the former case the structure is very fine, consisting of small grains of austenite in a ferrite matrix (*see* Fig. 17). Before equilibrium is reached the average composition of the austenite will be somewhere between those given by the A_3 and Ac_m lines; this corresponds to a greater amount of ferrite than that at equilibrium.

A further consideration is that equilibrium diagrams are strictly valid only for plane phase boundaries, as pointed out by Zener.[14] The determination of equilibrium diagrams usually involves slow heating or cooling, and therefore large particles. In the present case the total austenite–ferrite interface will be comparatively great because of the fineness of the structure (*cf.* Figs. 15*b* and 17), and an appreciable amount of energy might be required for the formation of these interfaces. Less energy will be available for the formation of austenite and therefore more ferrite will remain in the structure than would be expected from the equilibrium diagram.

As shown by Fig. 3, the onset of the pearlite formation leads to an increase in the amount of ferrite, even though the ferrite formation had apparently ceased. An explanation of this follows from a consideration of the equilibrium concentrations of carbon in ferrite in contact with either austenite or cementite. When the steady state is reached in the precipitation of ferrite, this ferrite may be assumed to be in equilibrium with the parent austenite and to have a carbon concentration as indicated by the extension of GP (the $\alpha - \alpha + \gamma$ phase boundary). It is therefore supersaturated with respect to cementite. The formation of pearlite starts by the production of a particle of cementite, most likely in the boundary between austenite and ferrite, and consequently a single grain of ferrite can be in contact with both cementite and austenite. In the vicinity of the cementite particle the carbon content of the ferrite decreases and a concentration gradient in the ferrite is set up. At 700° C. the diffusion coefficient of carbon in ferrite is more than one hundred times that of carbon in austenite.[15] As the further nucleation and growth of pearlite is slow just below A_1, there will be time for a transport of carbon from austenite to cementite through the ferrite. The formation of more ferrite will be accompanied by a thickening of the cementite particles. The thickening of cementite in contact with free ferrite is shown in Fig. 18.

As the transformation temperature is progressively lowered below A_1, the amount of ferrite in the final structure decreases, rapidly in the medium-carbon steels, owing to ever-earlier interruption by the formation of pearlite.

SUMMARY AND CONCLUSIONS

The isothermal formation of ferrite has been studied in a series of hypo-eutectoid plain carbon steels. The rate of formation of ferrite has been analysed in terms of nucleation and growth, with the following conclusions:

(i) At constant temperature the rate of nucleation and the rate of growth decrease with time, and so the rate of transformation decreases and a steady state in the reaction is approached.

(ii) With increasing degree of undercooling below the respective A_3 temperatures the rate of formation of ferrite increases, because both nucleation and growth rates increase. Whether they both pass through a maximum or not on further lowering of temperature is difficult to determine, but it does appear that the rate of growth reaches a maximum value for a finite degree of undercooling. The temperature dependence of nucleation is in qualitative agreement with theoretical expectations, based on calculations of the free-energy changes accompanying the formation of ferrite from austenite.

(iii) With decrease in carbon content the rate of formation of ferrite increases, owing to faster rates of nucleation and of growth. At a given level of undercooling the rate of growth is greater the lower the carbon content.

The steady-state amounts of ferrite increase with decrease in temperature to below the A_1 temperature, in agreement with metastable equilibrium conditions. When pearlite is formed after the steady state in ferrite formation has been reached, there is a further increase in the amount of ferrite. This is attributed to the solubility of carbon in ferrite in contact with austenite being different from that in ferrite in contact with cementite.

With further decrease in temperature below A_1 the formation of pearlite interrupts the formation of ferrite at an increasingly earlier stage, and so the composition of the final structure is governed by the kinetics of these two rival reactions.

Acknowledgments

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References

1. H. C. H. Carpenter and J. M. Robertson: *J. Iron Steel Inst.*, 1931, No. I, pp. 345–379.
2. E. S. Davenport: *Trans. Amer. Soc. Metals*, 1939, vol. 27, pp. 837–886.
3. "Atlas of Isothermal Transformation Diagrams": 1943, Pittsburgh, U.S. Steel Corporation.
4. J. R. Blanchard, R. M. Parke, and A. J. Herzig: *Trans. Amer. Soc. Metals*, 1941, vol. 29, pp. 317–335; 1943, vol. 31, pp. 849–863.
5. P. Payson, W. L. Hodapp, and J. Leeder: *Ibid.*, 1940, vol. 28, p. 314.
6. E. Greulich: *Arch. Eisenhüttenwesen*, 1945, vol. 18, pp. 173–184.
7. D. L. McBride, C. H. Herty, and R. F. Mehl: *Trans. Amer. Soc. Metals*, 1936, vol. 24, pp. 281–311.
8. H. Lange: *Mitt. Kaiser-Wilhelm-Inst. Eisenforsch.*, 1940, vol. 22, pp. 229–240.
9. R. F. Mehl: *J. Iron Steel Inst.*, 1948, vol. 159, pp. 113–129.
10. R. F. Mehl and A. Dubé: "Phase Transformations in Solids," pp. 545–582: 1951, New York, John Wiley and Sons, Inc.
11. E. S. Davenport and E. C. Bain: *Trans. Amer. Inst. Min. Met. Eng.*, 1930, vol. 90, pp. 117–144.
12. T. G. Digges: *J. Res. Nat. Bur. Stand.*, 1938, vol. 20, pp. 571–587.
13. R. F. Mehl and C. Wells: *Trans. Amer. Inst. Min. Met. Eng.*, 1937, vol. 125, pp. 429–469.
14. C. Zener: *Metals Technology*, 1946, vol. 13, Tech. Publ. No. 1925.
15. J. K. Stanley: *Trans. Amer. Inst. Min. Met. Eng.*, 1949, vol. 185, pp. 752–761; *J. Met.*, 1949, vol. 1, Oct.
16. E. Scheil: *Z. Metallkunde*, 1935, vol. 27, pp. 199–208.
17. F. C. Hull, R. A. Colton, and R. F. Mehl: *Trans. Amer. Inst. Min. Met. Eng.*, 1942, vol. 150, pp. 185–207; *Metals Technology*, 1942, vol. 9, Tech. Publ. No. 1460.
18. W. A. Johnson and R. F. Mehl: *Ibid.*, 1939, vol. 135, pp. 416–442.
19. C. Zener: *J. Appl. Phys.*, 1949, vol. 20, pp. 950–953.
20. C. A. Wert: *Ibid.*, 1949, vol. 20, pp. 943–949.
21. C. H. Johansson: *Arch. Eisenhüttenwesen*, 1937, vol. 11, pp. 241–251.
22. J. O. Clayton and W. F. Giauque: *J. Amer. Chem. Soc.*, 1932, vol. 54, p. 2623.
23. R. P. Smith: *Ibid.*, 1946, vol. 68, pp. 1163–1175.
24. C. Wells, W. Batz, and R. F. Mehl: *Trans. Amer. Inst. Min. Met. Eng.*, 1950, vol. 188, pp. 553–560; *J. Met.*, 1950, vol. 2, Mar.

Effect of Rust and Environment on Inhibition by Zinc-Rich Paints

By J. E. O. Mayne, D.Sc., F.R.I.C., A.R.C.S., D.I.C.

SYNOPSIS

In a previous publication it was shown that paint films containing 96% of zinc dust were capable of affording cathodic protection to pickled steel surfaces when immersed in sea-water. The present paper shows that these films can also protect rusty surfaces against corrosion by sea-water or rain-water. Rusty surfaces vary and so this conclusion should not be applied to all types of rusty surface.

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In a previous publication[1] it was shown that paints pigmented with zinc dust could confer cathodic protection on steel, provided that three conditions were fulfilled :

(i) The particles of zinc dust must be in metallic (*i.e.*, electronic) contact with each other
(ii) The zinc must be in metallic contact with the coated steel
(iii) There must be an electrolytic conductor (*e.g.*, water containing a trace of salt or acid) to complete the circuit between the two dissimilar metals.

It was established that the first condition was fulfilled when the zinc content of the dried paint film was of the order of 96 wt.-%, the second condition when pickled steel specimens were used, and the third when they were immersed in sea-water.

The following investigation was carried out to ascertain whether the zinc dust could make metallic contact with the steel through a layer of rust when zinc-rich paints were applied to rusty surfaces, rust being a poor electronic conductor. A further object was to find out whether, under conditions of atmospheric exposure where the concentration of electrolyte may be low, sufficient electrolyte was present to complete the circuit between the two metals and consequently allow the paints to function satisfactorily.

EXPERIMENTAL DETAILS

Method

To obtain information about the degree of contact between the zinc dust in the paint film and the steel, pickled and rusty specimens were painted and immersed in sea-water, rain-water, and distilled water and their potentials were measured over a period. Further information was obtained by measuring the potentials of painted specimens, which had been exposed on the roof.

Potential measurements afford a convenient method of assessing the degree of metallic contact between the zinc particles in the paint film and the steel, since when the contact is good the specimen will have the potential of a piece of zinc. On the other hand, if the electronic contact between the zinc and the steel is bad, the specimen will have the potential of uncoated steel.

Materials

The 3 in. × 1 in. steel specimens were cut from 18 S.W.G. hot-rolled mild-steel strip, the composition of which has been published elsewhere.[2] The standard laboratory grade zinc dust contained about 4% of zinc oxide and the particles were spherical, average size 5μ. The zinc-rich paint was prepared by mixing 500 g. of zinc dust with 100 g. of polystyrene lacquer, the composition of which has been published;[3] the mixture, contained in a small tin, was rotated on the rollers of a ball mill until the paint was free from lumps.

Marine Exposure

The partial immersion tests in sea-water were performed with the surface of the steel in three different conditions :

(i) *Pickled*—The surface mill scale was removed with dilute hydrochloric acid, containing a pickling inhibitor. The specimens were then washed under the tap, dried on filter paper, and stored in a desiccator
(ii) *Rusty*—The specimens were weathered on the roof for about 1 year, which removed all the mill scale
(iii) *Rusty and wire-brushed*—After weathering on the roof for 1 year the loose rust was removed with a mechanical wire brush.

Specimens were coated with zinc-rich paint to within ½ in. of the top ; they were allowed to dry for one week at room temperature and a brass terminal was attached to the unpainted portion. Care was taken that the brass did not make contact with the paint film ; in the case of the rusty specimens, a portion of the rust was removed with an emery wheel before the terminal was attached.

The painted specimens, together with a range of unpainted specimens and a piece of electrolytic zinc, were partially immersed in natural sea-water contained in a number of small jars. These were stored under a bell jar, with a water seal, in a thermostat maintained at 25° C. To make a measurement the sea-water was connected to a saturated calomel electrode, *via* a beaker of sea-water, by means of damp strips of filter paper and the potential of the system was measured from time to time with a valve potentiometer, the potential first being measured ½ hr. after immersion. The results obtained over 64 days are shown graphically in Fig. 1. The experiment

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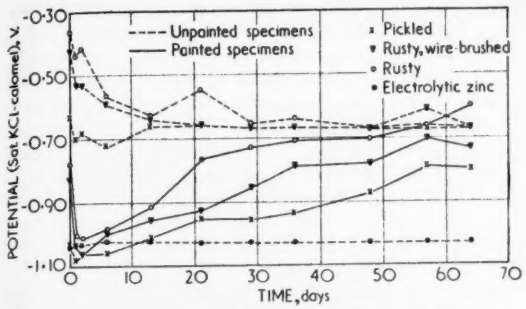

Fig. 1—Specimens with various surface conditions, painted and partially immersed in natural sea-water

was continued for 15 months ; over the period 2–15 months the potentials of all the specimens were of the same order. Throughout the whole period all the specimens were completely protected; in the case of the rusty specimens slight breakdown occurred at one or two places, but there was no rust in the sea-water. After 15 months' partial immersion the painted pickled specimen was removed and the paint film was broken by a scratch, which exposed the steel ; upon re-immersion, rust was formed at the scratch within 24 hr., but rusting soon ceased owing to stifling and it did not undermine the paint in the vicinity of the scratch line. It should be noted that when the scratch mark was made before immersion in sea-water no rust was visible at the scratch after 20 months.

Atmospheric Exposure

Two pickled specimens were coated with zinc-rich paint, dried, and partially immersed in rain-water and distilled water. The potential of each specimen was then measured over a period of time, the first readings being taken 1 hr. after immersion. This

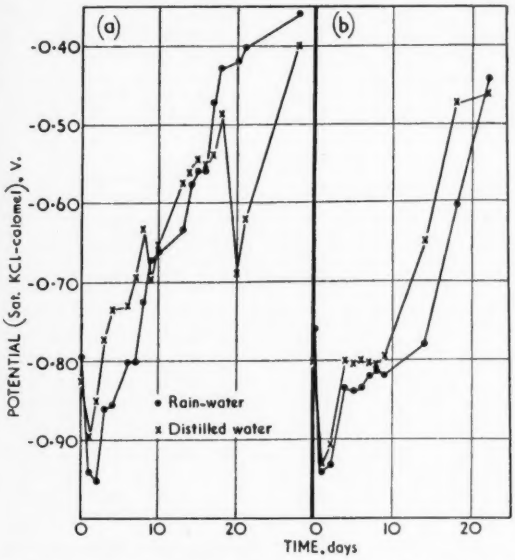

Fig. 2—(*a*) Painted pickled specimens and (*b*) painted, rusty, and wire-brushed specimens immersed in rain-water and distilled water

Table I

POTENTIALS OF PAINTED SPECIMENS EXPOSED ON ROOF

Potentials are expressed in millivolts and were all negative with respect to a HgCl,KCl (saturated) electrode

Time Exposed, days	Pickled	Rusty	50% Scale	Weathered 18 Days	Scale Intact
1	851	784	820	850	828
2	860	998	810	868	867
3	857	698	806	804	815
5	858	702	853	884	908
6	860	715	800	820	820
7	898	862	834	858	796
26	810	660	723	715	742

experiment was next repeated with two rusty specimens, which had been weathered on the roof for at least one year, the loose rust being removed by a mechanical wire brush just before painting. The results obtained are shown graphically in Fig. 2.

To obtain information about the effects of rust and broken scale on the contact between the zinc and the steel, particularly under conditions of atmospheric exposure where the specimens will be alternately wet and dry, further experiments were carried out in which painted specimens, exposed on the roof, were brought in from time to time and their potentials determined immediately after partial immersion in rain-water. In these experiments the steel was in five different surface conditions before painting :

(i) Pickled
(ii) Weathered until all the scale had gone, and loose rust removed with a mechanical wire brush
(iii) Weathered until half the scale had been removed, then wire-brushed
(iv) Weathered for 18 days, then wire-brushed
(v) Covered with intact mill scale.

The results obtained are shown in Table I.

In another series of experiments pickled specimens were allowed to weather on the roof, and at intervals a specimen was brought in, brushed with a mechanical wire brush, and coated with one or two coats of zinc-rich polystyrene paint ; when dry it was re-exposed. Further details of these exposure trials are to be found in another paper.[4] In one experiment, which had been in progress for 8 months, measurements were made of the potentials of all the specimens after 45 min. immersion in rain-water ; the results are shown in Table II.

DISCUSSION

Mechanism of Cathodic Protection

The corrosion of iron consists of the passage of iron ions from the metallic lattice into solution with the liberation of electrons, which are consumed elsewhere by reaction with water and oxygen. If the iron receives a supply of electrons, *i.e.*, is made more negative or cathodic, it becomes difficult for the positively charged iron ions to leave the metallic lattice and corrosion may be suppressed. In the presence of oxygen the electrons may be discharged in accordance with the equation

$$2H_2O + O_2 + 4e \rightarrow 4OH'.$$

Thus the *p*H at the surface of the iron may rise and as a result either the iron will be rendered passive owing to the formation of an invisible film, mainly

Table II

EFFECT OF WEATHERING ON THE POTENTIALS OF PAINTED SPECIMENS

Potentials are expressed in millivolts and were all negative with respect to a HgCl/KCl (saturated) electrode

Condition of Specimen before Painting	Period of Exposure after Painting, months	One-Coat System	Two-Coat System
Pickled	7	930	912
Weathered 1 month	7	860	812
Weathered 2 months	6	888	887
Weathered 3 months	5	871	910
Weathered 4 months	4	850	843
Weathered 5 months	3	941	800
Weathered 6 months	2	842	834

composed of Fe_3O_4 or γFe_2O_3,[5] or if divalent cations are present it may become coated with a visible deposit of an insoluble hydroxide or carbonate, which in time may be converted into a basic chloride. This formation of a protective film is of great importance in cathodic protection, particularly when the current is supplied from an external source, since it allows the current density to be decreased slowly without protection breaking down.

Protection by Zinc-Rich Paints

Zinc-rich polystyrene paints have been formulated so that the zinc particles in the dried film are in metallic contact with each other ; the dried films have appreciable electronic resistance and are porous, since the amount of vehicle has been reduced to a minimum ; furthermore, the zinc dust provides a source of divalent ions at the surface of the steel, *i.e.*, at the point at which they are most required.

The changes in potential of painted specimens over a period, when partially immersed in sea-water, rain-water, and distilled water, are shown in Figs. 1 and 2. All the curves have a similar shape, the potential showing a fall to a minimum value after 1–2 days' immersion and then rising steadily with time.

When iron is placed in metallic contact with zinc, electrons will flow from the zinc to the iron, the potential of which will be depressed, provided that the electric circuit between the two metals is completed by an electrolytic conductor. Since the films of zinc-rich paint are porous, oxygen will be present at the surface of the iron, which consequently will become alkaline and coated with a protective substance. The initial drop in potential can be attributed to the fact that this protective substance covers the steel formerly exposed at pores, giving the simple potential of zinc instead of the compromise potential of a steel/zinc couple.

In all the immersion experiments the minimum potential developed after 1–2 days' immersion. Zinc immersed in sea-water took up a potential of −1·020 to −1·030 V., whereas coated steel specimens fell to −1·080 V. when pickled, −1·060 V. when rusty and wire-brushed, and −1·009 V. when loose rust was present. According to Gilbert[6] the potential of zinc immersed in supply waters is −0·68 V. on the hydrogen scale, which corresponds to −0·93 V. on the saturated calomel scale. In the experiments with rain-water slightly more negative potentials were obtained, both with pickled and wire-brushed rusty surfaces. Thus, since all the rusty specimens used in these experiments acquired the potential of zinc after 1–2 days' immersion, it follows that zinc-rich paints can make metallic contact with the underlying metal through a layer of rust. This contact may be due to the penetration of the rust film at a few points by the zinc dust, the particles of which had a diameter of 5μ. Unfortunately little is known about the permeability of rust films, but if the view is taken that they are too continuous to allow the paint to function in this way, it is necessary to introduce the hypothesis that the rust is reduced with the formation of conducting bridges through the rust layer.

The most probable explanation of the ensuing steady rise in potential is that the zinc dust becomes attacked and the metallic contact between the particles is destroyed. The important point is that the paint does not then fail because the steel is now protected by a layer of hydroxide or carbonate, which has been formed in physical contact with the metal. This explanation is in accordance with the observation that a pickled specimen, freshly coated with zinc-rich paint, is protected at a scratch line upon immersion in sea-water, but if the specimen is first immersed in sea-water for 1 month and then scratched, upon re-immersion the scratch will block itself with rust within 24 hr.

So far, the behaviour of the paint has been examined only under immersed conditions. A few further experiments were therefore performed to determine the effects of atmospheric exposure, in which the paint film is alternately wet and dry. In these experiments steel in five different surface conditions was painted and exposed on the roof. From time to time specimens were brought in, immersed in rain-water, and their potentials measured. The results obtained are shown in Table I, from which it can be seen that good contact was established when the surface was pickled or carried mill scale, but with a rusty surface the results were erratic. In another experiment the effect of weathering pickled surfaces up to 6 months before painting was examined ; the results obtained are shown in Table II, from which it can be seen that good contact was made through the rust film on all the specimens.

CONCLUSIONS

The general conclusions which arise from these laboratory experiments are that zinc-rich polystyrene paints are capable of affording cathodic protection to both pickled and rusty steel when immersed in sea-water, rain-water, and distilled water. With pickled surfaces the contact between the zinc and the steel was good ; on rusty surfaces contact took place at a sufficient number of points for the paint to function, but probably less efficiently. Furthermore, atmospheric exposure tests have indicated that the paints can function, under normal conditions, when applied to pickled surfaces and that they may function when applied over broken mill scale and rust. Rusty surfaces vary and it is clear that in some cases the rust may be so continuous that it will prevent efficient contact between the zinc in the paint film and the steel ; in these circumstances the paint film will be undermined and may flake off rapidly. Consequently, it is important to know just how much rust or broken

mill scale a surface can carry before it becomes unsuitable for protection by zinc-rich paints. Exposure trials have been carried out in Cambridge designed to provide information regarding this problem.

Acknowledgments

This investigation formed part of a programme of research carried out for the British Iron and Steel Research Association by the group of investigators associated with Dr. U. R. Evans. The author is indebted to B.I.S.R.A. and Dr. U. R. Evans for their support, and wishes to thank Dr. J. N. Agar for constructive criticism and Mr. C. Taylor for help with the measurements.

References

1. J. E. O. MAYNE : *Journal of the Society of Chemical Industry*, 1947, vol. 66, pp. 93–95.
2. J. E. O. MAYNE and R. S. THORNHILL : *J. Iron Steel Inst.*, 1948, vol. 158, p. 228.
3. J. E. O. MAYNE : *Journal of the Oil and Colour Chemists' Association*, 1949, vol. 32, p. 483.
4. J. E. O. MAYNE : *J. Iron Steel Inst.*, 1954, vol. 176, pp. 143–146 (this issue).
5. J. E. O. MAYNE, J. W. MENTER, and M. J. PRYOR : *J. Chem. Soc.*, 1950, pp. 3229–3236.
6. P. T. GILBERT : *Sheet Metal Ind.*, 1948, vol. 25, p. 2442.

Atmospheric Exposure Tests with Zinc-Rich Polystyrene Paints

By J. E. O. Mayne, D.Sc., F.R.I.C., A.R.C.S., D.I.C.

SYNOPSIS

Atmospheric exposure tests in Cambridge have indicated that single coats of zinc-rich polystyrene paint can protect rusty surfaces, since specimens were in good condition after nearly 4 years' exposure, provided that the rust was not more than 6 months old. Variable results were obtained when the rust was older, but better results were obtained when painting was carried out in June than in December; thus, in one experiment specimens weathered 26–27 months before painting were in good condition after 4½ years' subsequent exposure.

Steel specimens initially carrying mill scale and weathered for short periods tended to break down locally, but after 9 months' weathering they behaved similarly to weathered pickled specimens.

The use of zinc-rich polystyrene paint as a primer has been examined. The best results were obtained when a sealing coat of an iron-oxide linseed-oil paint was used; the results were slightly better than those obtained with a red-lead linseed-oil primer. With all the paint systems good protection was obtained on rusty surfaces only when painting was done in the summer. This seasonal variation is attributed to atmospheric pollution, and suggestions are made for future work. 848

IT has been shown* that under atmospheric conditions zinc-rich polystyrene paints were capable of giving cathodic protection to steel, even when applied over rust or broken mill scale. The object of this investigation was to obtain information about the amount of rust or broken mill scale that a steel surface could carry before it became unsuitable for protection by zinc-rich polystyrene paints.

The experimental procedure consisted in exposing, on the roof of the Cambridge laboratory, pickled steel specimens, and others carrying intact mill scale; at intervals specimens were brought in, wire-brushed with a mechanical wire brush, and coated with zinc-rich polystyrene paint; when dry they were re-exposed. Thus, in each experiment the period of weathering plus the time the painted specimens were exposed was constant, since it represented the duration of the experiment.

The behaviour of single and double coats of zinc-rich polystyrene paint has been examined and the use of these paints as primers under coats of iron oxide paints has been investigated, a standard red-lead linseed-oil priming paint being used for comparison. In one series of tests a zinc dust containing a certain proportion of large particles was used and in another a portion of the zinc dust was replaced by graphite; the effect of metallic lead powder in the paint was also examined. No improvement was observed as a result of these modifications. In all, nine series of experiments have been carried out, and these can be divided into two groups; the first has been in progress for 6–7 years and the second for nearly 4 years.

EXPERIMENTAL DETAILS

Materials

The 3 in. × 2 in. steel specimens were of the same material as those used in a previous investigation.* All the specimens were degreased with carbon tetrachloride and were exposed either in the as-rolled condition or after pickling in dilute hydrochloric acid containing a pickling inhibitor.

The zinc dust was the same as that used in the previous investigation.* Two experiments were carried out with a special grade of zinc dust, the particles of which were irregular and had an average size of 20 × 50μ. This zinc dust was selected from a number of samples, because, when used in a paint, the dried film had a much lower resistance than a similar film prepared from the standard laboratory grade. The

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*J. E. O. Mayne, *J. Iron Steel Inst.*, 1954, vol. 176, pp. 140–143 (this issue).

compositions and methods of manufacture of the paints are given in the Appendix.

Behaviour of Single Coats of Zinc-Rich Polystyrene Paints

Series 1 *and* 2—Twelve pickled specimens and a similar number carrying intact mill scale were exposed on the roof. After a certain period a specimen was withdrawn, wire-brushed with a mechanical wire brush, and weighed. The electrical conductivity of the corrosion products was then examined, using an apparatus consisting of a dry battery and a galvanometer connected in series with a spring clip and a blunt brass electrode, made from $\frac{1}{4}$-in. brass rod with a spherical end. To carry out a test the spring clip was attached to one corner of the specimen and the brass electrode was passed lightly over its surface. The area covered with mill scale was then estimated, and the specimens were painted, dried, re-weighed, and re-exposed. It was found that in the case of pickled specimens the surface became non-conducting after 49 days' exposure. The surface of the specimens which initially carried mill scale remained conducting at those points where the scale remained, but not at the rusty areas ; consequently, the surface became non-conducting only after all the mill scale had disappeared, after about 9 months' exposure.

All the specimens were examined after a total period of exposure of 3 years and again after 3 years 8 months ; their final condition and other relevant details are shown in Table I. Figure 1 shows all the specimens after exposure for 3 years 8 months.

Zinc-rich polystyrene paint applied to rusty pickled specimens was in good condition after nearly 4 years' exposure, provided that the rust was not more than 6–7 months old. When specimens carrying mill scale were used the results were erratic, owing to the tendency of the mill scale to be undermined.

There is a striking similarity between the pairs of specimens (*i.e.*, pickled and scale intact) weathered for 256 days or longer, the specimen weathered longest being in the best condition. This cannot be due to the fact that this paint film has had the shortest exposure, because 8 months after the last examination there had been little change in the condition of the specimens.

These results suggest that rust undergoes cyclic variations in its suitability for painting. Experiments of longer duration have confirmed this view.

Series 3–5—These experiments were carried out in a similar way to those in series 1 and 2, but differed in that the exposure was for a longer period and that a zinc-rich polystyrene paint prepared from normal zinc dust was compared with one made from a zinc dust having a certain fraction of large particles, so that the dried paint film had a higher conductivity.

The condition of the specimens after 80 months' exposure is shown graphically in Fig. 2, in which the percentage of paint which has failed (*i.e.*, lifted or disappeared) is plotted against the number of months the specimens were weathered before being wire-brushed and painted. The weight of paint applied is also given for each specimen ; six specimens were given in error two coats of paint, and the weights of paint applied were recorded only after the experiment had been in progress for 4 months.

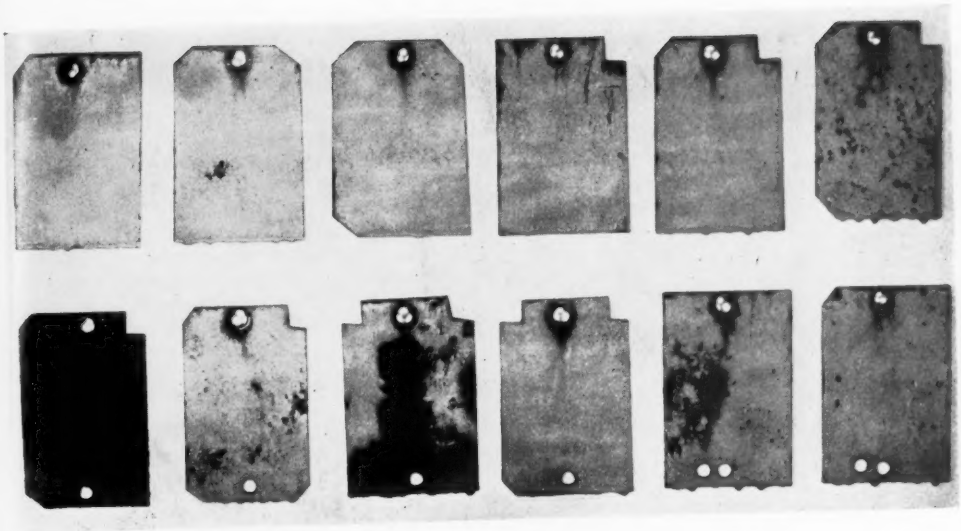
The similar shape of all three curves in Fig. 2 suggests that there was very little difference between the two grades of zinc dust, and that weathered specimens undergo cyclic variations in their suitability for painting. Thus, in all three cases the specimens weathered for 20–22 months failed completely, but specimens

Table I

CONDITION OF SPECIMENS AFTER 3 YEARS 8 MONTHS

Specimens had surface area of 12 sq. in.

Period of Weathering, days	Weathered Pickled Specimens		Weathered Specimens (Initially Carrying Mill Scale)		
	Weight of Paint Applied, g.	Condition after 3 years 8 months	Weight of Paint Applied, g.	Estimated Area of Intact Scale before Painting, %	Condition after 3 years 8 months
11	2·9	Good, but corrosion starting at top edge	2·8	20	All paint flaked off
31	4·7	Good, blistered at one point, some corrosion at edges	0·9	10	10% failed by blistering
49	...	Good, corrosion starting at edges	...	10	70% failed by flaking
70	2·1	Good, corrosion starting at top edge	2·9	4	Good, corrosion starting at edges
106	2·4	Good, corrosion starting at top edge	2·6	25	25% failed by flaking
139	3·1	10–20% failed, paint starting to lift	2·7	15	Good, corrosion starting at numerous pinpoints
192	2·2	Good, corrosion starting at edges and at a number of pinpoints	2·9	10	Good, corrosion starting at numerous pinpoints
256	4·9	50% failed, paint badly cracked and starting to lift	5·9	Nil	50% failed, paint badly cracked
316	2·6	All paint flaked off	2·7	Nil	All paint flaked off
368	5·2	70% failed, remaining paint flaking	4·0	Nil	70% failed, rest flaking
407	4·9	Paint badly cracked and on point of flaking	4·4	Nil	Paint badly cracked and on the point of flaking off
503	3·8	Paint starting to lift at various points	2·9	Nil	Paint has started to lift at a few points



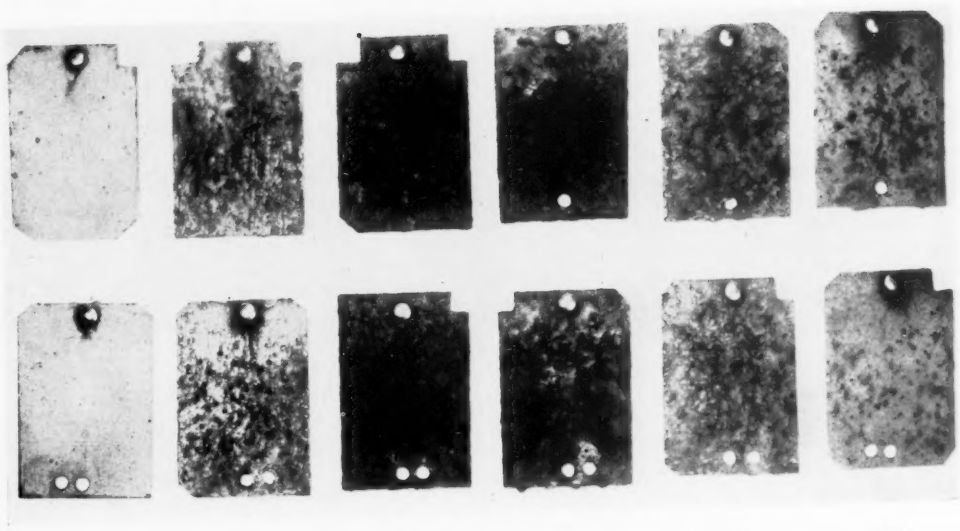


Fig. 1—Condition of specimens painted with zinc-rich polystyrene paint after exposure for 3 years 8 months on the roof of the Cambridge laboratory

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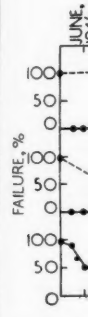

Fig. 2—E
zinc-

weathered for shorter or longer periods were in better condition. This result was not due to variations in paint thickness, but there is some indication that it is associated with atmospheric pollution.

Series 6–8—These experiments were designed to find out whether the addition of graphite or metallic lead powder to a zinc-rich polystyrene paint was advantageous. Three paints were prepared : the first was a normal zinc-rich polystyrene paint ; in the second 3 wt.-% of the zinc dust was replaced by graphite ; in the third 40 wt.-% of the zinc dust was replaced by metallic lead powder. The experimental procedure was the same as that used previously, specimens carrying intact mill scale being used.

The experiments have now been in progress for more than 7 years, and they indicate that no advantage is to be gained by the addition of either graphite or lead powder to a zinc-rich polystyrene paint.

Behaviour of Zinc-Rich Polystyrene Paints in Two-Coat Systems

Series 9—The object of this series of experiments was to examine the use of zinc-rich paints as primers on steel carrying intact mill scale which had been weathered for various periods. Two finishing paints were used : iron oxide dispersed in a polystyrene lacquer and iron oxide ground in linseed oil. A standard red-lead priming paint was included in the experiments and the behaviour of specimens carrying two coats of zinc-rich polystyrene was also examined.

Five two-coat systems were used :

(i) Zinc + polystyrene/zinc + polystyrene
(ii) Zinc + polystyrene/iron oxide + polystyrene
(iii) Zinc + polystyrene/iron oxide + linseed oil
(iv) Red lead + linseed oil/iron oxide + linseed oil
(v) Red lead + linseed oil/iron oxide + polystyrene.

The experimental procedure was the same as that used previously ; the compositions of the paints are given in the Appendix.

The condition of the specimens after 89 months' exposure is shown graphically in Fig. 3. No account has been taken of the darkening of the paint around the fixing screw or at the edges.

When two coats of zinc-rich polystyrene were used the protection afforded was not so good as when one coat of zinc-rich polystyrene was followed by a sealing coat. Both the sealing coats examined contained iron oxide, but slightly better protection was obtained

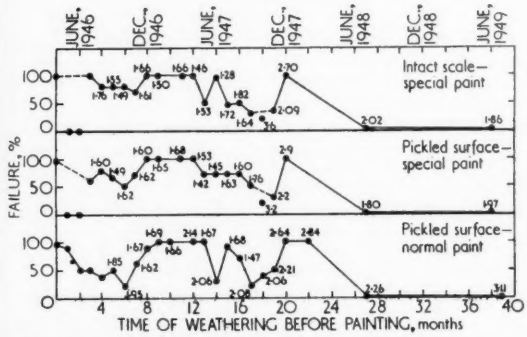

Fig. 2—Behaviour of single coats of normal and special zinc-rich paints weathered for different periods

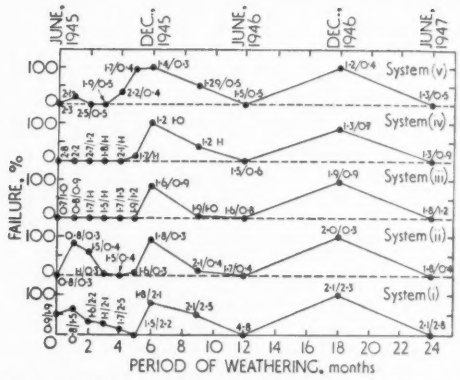

Fig. 3—Behaviour of two coat paint systems on surfaces weathered for different periods

when the pigment was dispersed in linseed oil than when polystyrene was used.

A similar result was obtained when the priming paint consisted of red lead+linseed oil. Iron oxide+polystyrene did not adhere well when applied over red lead+linseed oil, and in all cases the sealing coat showed signs of flaking ; the surfaces of all the specimens were marked with small circular depressions where the sealing coat had flaked away.

The best two systems examined were (iii) and (iv). The behaviour of both systems was similar but, if darkening of the paint around the edges and at the fixing screw is neglected, there is some indication that zinc-rich polystyrene has given a slightly better performance than red lead+linseed oil, when used as a primer, since the colour of the specimens is lighter and in several instances the corrosion is slightly less.

Figure 3 shows that throughout these experiments there was a cyclic variation in the behaviour of all the paints examined. This variation appears to be associated with the time of year at which painting took place ; specimens painted in December 1945 and 1946 were all in bad condition, whereas those painted in June 1945, 1946, and 1947 were almost perfect. The effect cannot be attributed to variations in the time of exposure of the paint film, since very little change has occurred in the relative condition of all the specimens during the last 8 months.

CONCLUSIONS

Single Coats of Zinc-Rich Polystyrene Paint

Single coats of zinc-rich polystyrene paint applied to a pickled surface and to weathered pickled surfaces were in good condition after nearly 4 years' exposure on the roof at Cambridge, provided that the surface was not weathered for more than about 6 months before painting. When weathered for longer periods the results were variable, but specimens painted in June, 1948, which had been weathered for 26–27 months before painting, were in good condition after 4½ years' exposure. Zinc-rich paints can give effective protection to rusty steel under atmospheric conditions, provided that the rust, before painting, is in a suitable condition.

Steel specimens initially carrying mill scale and

weathered for short periods tended to break down locally, but after they had been weathered for about 9 months they behaved in a similar way to weathered pickled specimens. This local breakdown was probably associated with the presence on the surface of the specimens of residual mill scale, which was removed by 9 months' weathering.

Zinc-Rich Polystyrene Paints as Priming Coats

The best of the five two-coat paint systems were :

Zinc + polystyrene/iron oxide + linseed oil
Red lead + linseed oil/iron oxide + linseed oil.

Exposure trials of these two systems have now been in progress for nearly 7½ years and in several cases the specimens are in good condition. Very little difference has been observed in the behaviour of the two systems, but there is a slight indication that zinc-rich polystyrene is a little better than red lead + linseed oil when used as a primer.

Seasonal Changes in the Nature of Rust

Throughout this investigation there has been an indication that rusty specimens painted in June have a longer life than those painted in December. This observation has been made both with single coats of zinc-rich polystyrene paint and with a number of two-coat systems. The effect cannot be attributed to seasonal variations in the drying of unsaturated natural oils, since it has been obtained with polystyrene-based paints, which dry by solvent evaporation ; neither can it be due to the presence of moisture, since previous work in these laboratories* has indicated that steel surfaces painted wet do not break down more rapidly than similar surfaces painted dry, when drying oil paints are used. The most probable explanation of this seasonal variation lies in the atmospheric pollution. It has been known for some time that the atmosphere in Cambridge is rural in summer but industrial in winter, owing to the combustion of coal in a large number of open grates.

Atmospheric pollution may accelerate the breakdown of the paint film. However, it has been found that specimens painted in June behave better than those painted in the following December, and so this acceleration of breakdown can only occur when the paint is fresh and not more than 6 months old ; furthermore, the cyclic variation has been obtained with two very different types of paint.

It is difficult to see how a paint film can become more resistant after some months' exposure, but this difficulty does not exist if the view is accepted that the life of a paint film on a rusty surface is affected by the presence in the rust of products arising from atmospheric pollution. The improved results, obtained when painting took place in June, can then be attributed to the removal of the detrimental products by the leaching action of the spring rains.

If this explanation is correct, the following suggestions deserve consideration. It is impossible to weather industrial structures in a rural atmosphere before the application of paint, but it might be possible to remove or nullify the detrimental products by a washing treatment, such as the use of a mechanical wire brush in a stream of fresh water, with or without the addition of some cheap chemical.

If more was known of the part played by atmospheric pollution in the atmospheric corrosion of iron, it should be possible to devise some simple test to decide whether a rusty surface was in a good or bad condition for painting. It is suggested that future work should be directed along these lines, since it is possible that the life of a paint coat applied to a rusty surface could be prolonged by some simple treatment or series of treatments before painting.

* J. E. O. Mayne and U. R. Evans, *Journal of the Society of Chemical Industry*, 1949, vol. 68, pp. 212–215.

Acknowledgments

This investigation formed part of a programme of research carried out for the British Iron and Steel Research Association by the group of investigators associated with Dr. U. R. Evans. The author is indebted to B.I.S.R.A. and Dr. U. R. Evans for their support ; he also wishes to thank Mr. C. Taylor for help in the preparation of the specimens and Dr. H. Mills, of the Imperial Smelting Corporation Ltd., for supplying a number of samples of zinc dust.

APPENDIX

Preparation of the Paints

Zinc-Rich Polystyrene

Polystyrene O.P.S.3	15 g.
Aroclor 1254	5 g.
Xylene	80 g.
Zinc dust	500 g.

The polystyrene was added to the Aroclor and xylene contained in a bottle, which was slowly rotated on the rollers of a ball mill until solution was complete. The zinc dust was then added and the rotation continued until the paint was free from lumps.

Iron Oxide + Polystyrene

Polystyrene O.P.S.3	15 g.
Xylene	80 g.
Tricresyl phosphate	5 g.
Iron oxide	10 g.

The tricresyl phosphate and iron oxide were ground on a triple-roll mill until free from lumps ; the resulting paste was then added to the lacquer prepared by dissolving the polystyrene in the xylene.

Iron Oxide + Linseed Oil

Medium stand oil	20 g.
Alkali refined linseed oil	40 g.
White spirit	30 g.
Iron oxide	120 g.

The stand oil and the linseed oil contained 0·4% of lead and 0·08% of manganese in the form of naphthenates. The iron oxide was ground with a portion of the drying oils on a triple-roll mill until free from lumps ; the remaining oil was then added and the mixture was thinned with the white spirit.

Red Lead + Linseed Oil

Medium stand oil	10 g.
Alkali refined linseed oil	20 g.
White spirit	10 g.
Red lead (non-setting)	160 g.

The stand oil and the linseed oil contained 0·4% of lead and 0·08% of manganese in the form of naphthenates. This paint was prepared in a similar way to the iron oxide + linseed oil paint.

Kinetics of First-Stage Graphitization in Iron–Carbon–Silicon Alloys

By J. Burke, B.Eng., Ph.D., and W. S. Owen, M.Eng., Ph.D., A.I.M.

SYNOPSIS

The reaction-rate curves for a series of high-purity iron–carbon–silicon alloys isothermally graphitized in the austenitic range have been determined dilatometrically. The kinetic data are satisfactorily explained in terms of an idealized model of the mechanism of graphite growth. The assumptions on which this analysis rests are examined in detail. The growth and nucleation rates of graphite have been measured by direct experiment. The results show that although the model is sufficiently accurate to provide a general picture and a basis for further study, it is not a very close approximation in certain details.

Activation energies for the overall reaction, nucleation, and growth have been measured and their interrelation is discussed. These studies throw new light on the processes controlling the rate of the reaction. 829

A HYPO-EUTECTIC iron–carbon–silicon alloy which has solidified in the metastable form at room temperature consists of massive cementite in a matrix of pearlite. The alloy may be transformed to the stable condition by a two-stage process:

(i) Conversion of the eutectic cementite to graphite and austenite by heating at temperatures above the Ac_1 point

(ii) Formation of a stable graphite–ferrite mixture either by very slow cooling through the eutectoid temperature range or by prolonged annealing slightly below this range.

The kinetics of these graphitization reactions in commercial white irons (the malleablizing process) have been extensively studied. In much of the work inadequate control has been exercised over experimental variables such as the arrangement of heat-treatment cycles (in some cases, no clear distinction has been made between the different stages of the transformation), the impurities or added elements, the melting and casting conditions, and the pre-graphitizing treatment. Thus, it is not surprising that some of the results are contradictory. As a basis for conjecture about the nature of the mechanisms involved in the reactions they have been productive but at times misleading. By restricting the present study to the first stage of graphitization in ternary alloys of high purity annealed under simple reproducible conditions, information has been obtained which provides a rational basis for the discussion of the kinetics of the malleablizing process.

A review and appraisement of the current ideas about the nature of the reaction in commercial alloys appears elsewhere.[1]

EXPERIMENTAL MATERIALS

The constituent elements used in the investigation were B.I.S.R.A. oxidized high-purity iron (A.G.L.2), silicon (total impurities 0·05%), and graphite (ash content 0·07%). The chemical analysis of these materials is given in Table I. The iron was first deoxidized by high-frequency melting in a pure-graphite crucible with the addition of a little powdered graphite. The resulting iron–carbon alloy was used to make up master iron–carbon and iron–carbon–silicon alloys. The ingots used in the experiments were produced by melting suitable proportions of the master alloys in glazed silica crucibles and casting into ¾-in. dia. water-cooled copper moulds. Every effort was made to ensure that the melting and casting conditions were the same in each case. Four of the ingots (melt 35) with constant carbon and controlled silicon contents were cast from a 600-g. split heat. The compositions of all the ingots are given in Table II.

Several sections of each 3-in. ingot bar were examined metallographically. All appeared to be homogeneous and free from inclusions.

MEASUREMENT OF REACTION RATES

Isothermal reaction-rate curves were determined by measuring length-changes in a dial-gauge vertical dilatometer, designed to allow the specimen to be heated *in vacuo*. To minimize the possible effect of small variations in grain size across the section each ingot was split longitudinally into quadrants, and the dilatometer specimens (⅝ in. long) were ground out of each quarter. The instrument was capable of detecting length-changes of 0·0001 in. A specially constructed electric-resistance furnace, controlled through Pt/Pt–Rh thermocouples suspended inside the furnace, was used to heat the specimen. At 1100° C. the temperature of the specimen did not vary by more than ± 2° C. This temperature was measured by means of a calibrated Pt/Pt–Rh thermocouple and a null-deflection precision potentiometer with an accuracy of about 1 μV. (corresponding to 0·1° C.).

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Dr. Burke, formerly a graduate student in the Department of Metallurgy, University of Liverpool, is now in the Research and Development Department Branch, Division of Atomic Energy, Salwick, Preston, and Dr. Owen is Lecturer in Metallurgy in the University of Liverpool.

Table I
COMPOSITION OF MATERIALS

Element	B.I.S.R.A. Iron	Silicon
Carbon, %	0·008	0·01
Silicon, %	0·002	Remainder
Manganese, %	0·005	—
Sulphur, %	0·006	—
Phosphorus, %	0·001	0·005
Nickel, %	0·010	—
Chromium, %	0·001	—
Copper, %	0·006	0·022
Aluminium, %	0·001	0·022
Cobalt, %	—	0·006
Iron, %	Remainder	0·0006
Titanium, %	—	0·003
Zirconium, %	—	0·003

Before starting each experiment the vacuum pumps and the furnace were allowed to run for several hours to establish equilibrium conditions. It took less than 5 min. to raise the temperature of the specimen from room temperature to the ambient furnace temperature. In general this was a much shorter time than the reaction incubation period, and so no serious error was introduced. However, the adoption of a convention for fixing zero time was necessary. In keeping with the usual practice, the time was measured from the instant the specimen was placed in the furnace.

The increase in length of the specimen, expressed as a fraction of the total increase in length for complete reaction, was taken to be a direct measure of the progress of the reaction. This fraction was plotted against the logarithm of the time. Ingot 19 was studied at six, 27 at five, 33 at four, and each of the four ingots from melt 35 at five temperatures in the austenitic range. All the curves were remarkably similar in their general form and could be superimposed simply by moving the zero of the time axis. Representative curves (from ingot 35/4) are shown in Fig. 1*a*. The curves obtained from ingots 35/1, 35/2, 35/3, and 35/4 at 888° C., showing the effect of increasing silicon content, are given in Fig. 1*b*.

The experimental data have been assembled in the form of isothermal transformation diagrams, the 'start' and 'end' of the reaction being taken as 5% and 95% transformation, respectively. For each alloy, the time for completion of the first stage of graphitization and the incubation period (on logarithmic scales) are approximately linear functions of temperature (° C.), and the two lines are approximately parallel. A typical diagram (for ingot 19) is shown in Fig. 2. In Fig. 3*b* the time for completion of the reaction in ingots 35/1, 35/2, 35/3, and 35/4 is plotted against temperature. Although variation in silicon content alters the position, it has no effect on the slope of the linear plot. Figure 3*a* shows that this is also true of the plot of incubation period against temperature. The slope of each plot made during the course of the investigation was measured and the average value was found to be $-0{\cdot}010$, the scatter being $\pm 0{\cdot}002$.

ANALYSIS OF ISOTHERMAL REACTION DATA

The experimental data can be conveniently examined by using a rate equation:

$$y = 1 - \exp\left(-\frac{t}{K}\right)^n \quad \text{.....(1)}$$

Table II
ANALYSIS OF INGOTS

Ingot Number	Carbon, %	Silicon, %
19	2·75	1·18
27	2·01	0·78
33	3·58	1·12
35/1	3·37	0·68
35/2	3·35	0·98
35/3	3·38	1·13
35/4	3·43	1·37

which, on differentiating, gives

$$\frac{dy}{dt} = \frac{n}{K^n}(1 - y)t^{n-1} \quad \text{.....(2)}$$

where y is the fraction transformed at time t, K is a temperature-dependent rate constant, and n is a constant.

Equation (1) is similar to an equation derived by Zener[2] to describe the kinetics of the growth of a precipitate in a supersaturated solid solution of infinite extent. According to Zener, in such a reaction

$$\frac{dy}{dt} = Bt^m \quad \text{.....(3)}$$

where B is a temperature-dependent rate constant and m is a constant.

In the simple case the index m depends only upon the shape of the precipitate. To account for the decrease in reaction speed towards the end of the process it is usual to assume that the particles of precipitate eventually interfere and that the resulting decrease in velocity is proportional to the fraction transformed. Thus, equation (3) becomes

$$\frac{dy}{dt} = B(1 - y)t^m \quad \text{.....(4)}$$

Although the introduction of the factor $(1 - y)$ may seem to be somewhat arbitrary, in a number of cases the experimental results appear to justify its adoption. For example, equation (4) has been applied successfully to data for the kinetics of the first, second, and third stages of tempering carbon steels by Cohen *et al.*[3, 4, 5] and to the precipitation of carbon and nitrogen from alpha iron by Wert.[6]

Johnson and Mehl[7] studied the kinetics of the eutectoid reaction in steels in terms of a model which assumed constant rates of nucleation N and of radial growth G, and a random distribution of spherical particles of precipitate. They derived the expression

$$y = 1 - \exp(-\tfrac{\pi}{3}NG^3t^4) \quad \text{.....(5)}$$

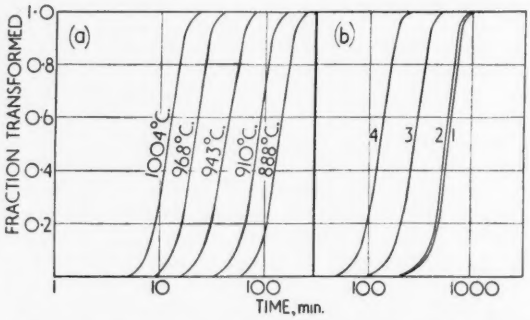


Fig. 1—Isothermal graphitization curves for (*a*) ingot 35/4 at various temperatures, (*b*) melt 35 at 888° C.

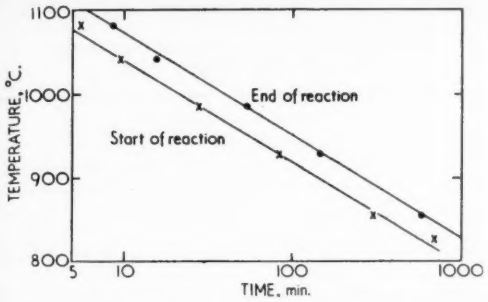


Fig. 2—Isothermal transformation diagram for melt 19

which leads to

$$\frac{dy}{dt} = A(1-y)t^3 \quad \ldots\ldots(6)$$

where

$$A = \frac{4}{3}\pi NG^3,$$

which is comparable with equation (4), when $m = 3$.

The index m is determined by the geometric form of the precipitate[2] and the kinetics of the nucleation process* and is assumed to be independent of temperature. Hollomon[8] has pointed out that it may be modified by special forms of impingement of the precipitate particles. However, these conditions are unlikely to be encountered in the present study. The constant B in equation (3) provides for the temperature-dependence of the reaction velocity and follows the usual exponential form:

$$B = C_1 \exp\left(-\frac{Q}{RT}\right) \quad \ldots\ldots(7)$$

where Q = activation energy for the reaction, cal./mole
T = absolute temperature, ° K.
C_1 = constant
R = gas constant.

It is usual to obtain values of activation energies from experimental rate constants with dimensions $(\text{time})^{-1}$. Zener[9] remarked on the fact that activation energies obtained from values of B, which has dimensions $(\text{time})^{-(m+1)}$, therefore require adjustment before being used for comparing the characteristics of different reactions.

The use of a rate equation of the form of equation (1) avoids this inconvenience. Moreover, it has the advantage that values of K can be obtained directly from the reaction curves in a simple way. Putting equation (1) in the logarithmic form,

$$\ln\left(\frac{1}{1-y}\right) = \left(\frac{t}{K}\right)^n \quad \ldots\ldots(8)$$

from which it can be seen that when $y = (e-1)/e$ (*i.e.*, approx. 0·63), $K = t$. Thus, the value of K at a particular temperature is given by the time at which $y = 0{\cdot}63$.

Since equation (1) may be written in the form

$$\log\log\left(\frac{1}{1-y}\right) = n\log t - n\log K - \log 2{\cdot}3 \quad \ldots\ldots(9)$$

the plot of $\log\log(1/1-y)$ against $\log t$ yields a straight line of gradient n. Excellent straight-line plots resulted when the experimental data obtained in the present study were analysed in this manner. Representative plots are shown in Fig. 4. The values of n for 33 of the reaction curves are listed in Table III. In this type of plot small differences in the slope are reflected as substantial differences in the numerical value of the gradient, and so it is not surprising that the results show some scatter. No definite trend in the values of n with variation in temperature or composition could be detected. The average value is 4·07.

The shape of curves conforming to equation (1) depends only upon the value of n. This is demonstrated in Fig. 5, in which hypothetical curves are plotted for different values of n and K. Thus, the

* The index m will only be a constant when the nucleation rate N can be expressed in the simple form $N = N't^a$, where N' and a are true constants.

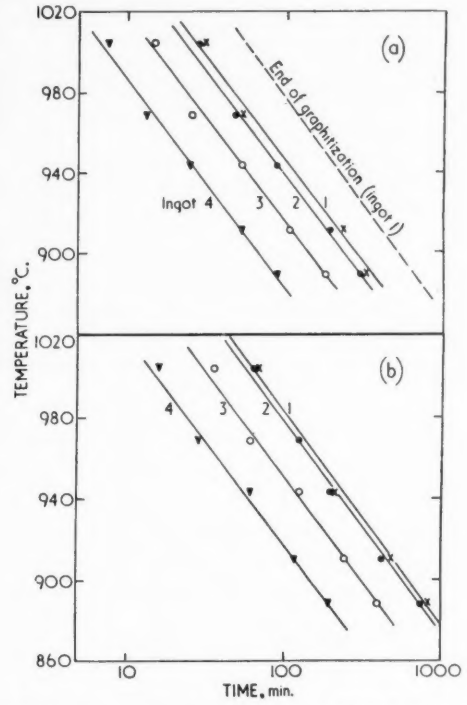


Fig. 3—Temperature *vs.* (*a*) start and (*b*) end of graphitization (melt 35)

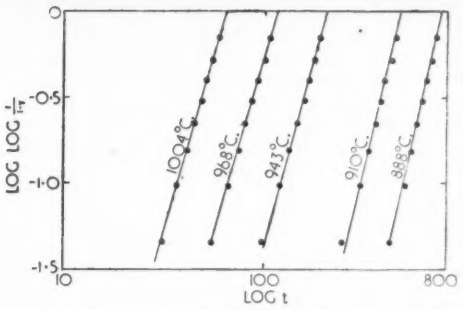


Fig. 4—Graph of $\log\log\left(\frac{1}{1-y}\right)$ *vs.* $\log t$ (for melt 35/1)

Table III

VALUES OF *n* FOR FIRST-STAGE GRAPHITIZATION

Temp., °C.	Ingot Number						
	19	27	33	35/1	35/2	35/3	35/4
786		5·01					
811			3·17				
828	4·40						
842		4·17					
853			2·86				
854	4·53						
888		4·30		5·46	5·28	3·90	3·91
900			3·30				
910				5·40	4·30	3·84	3·90
927	4·66						
938		4·26					
943				4·43	4·03	3·63	3·55
958		3·50					
968			3·28	4·15	3·84	3·54	3·99
984	4·54						
1004				4·20	3·91	3·46	3·87

Table IV

VALUES OF *K* FOR FIRST-STAGE GRAPHITIZATION, min.

Temp., °C.	Ingot Number						
	19	27	33	35/1	35/2	35/3	35/4
786		8600					
811			4300				
828	1170						
842		1750					
853			1040				
854	480						
888		310		640	600	315	147
900			245				
910				400	330	190	93
927	120						
938		89					
943				170	150	94	46
958		55					
968			70	95	86	47	23
984	43						
1004				53	51·7	28	13

fact that the experimental reaction curves can be superimposed is further evidence of the constancy of the value of n.

The experimental values of K are listed in Table IV, derived from

$$\frac{1}{K} = C_4 \exp\left(-\frac{Q_K}{RT}\right) \quad \ldots\ldots\ldots\ldots\ldots\ldots (10)$$

where Q_K cal./mole is the activation energy for first-stage graphitization. Hence Q_K can be evaluated by plotting the logarithm of K against $1/T$ and measuring the slope. Alternatively, the reciprocal of the absolute temperature may be plotted against the logarithm of the time t to a given fraction of transformation (*i.e.*, to constant y), say 50%. A typical plot is shown in Fig. 6. The values of Q_K in the seven ingots are listed in Table V. It is usual to consider that the values of activation energies determined in this manner have a probable error of $\pm$ 6000 cal./mole, and hence the divergence of the values in Table V is not serious. Within the limits of the experimental error, no trend of the value of Q_K with silicon content could be found. It was thought justifiable to average all the determinations to find the activation energy for first-stage graphitization, the result being 68,000 cal./mole.

Since the index n appears to be a characteristic of first-stage graphitization in iron–carbon–silicon alloys, its significance merits further consideration. The conditions postulated by Zener are not applicable to the present case, and so it is not surprising that the value of n (approx. 4) is not comparable with any value previously discussed. The experimental value of n can be explained in terms of an idealized model based upon the following assumptions (the validity of these will be discussed more fully later): (i) The growing graphite nodules are spherical; (ii) the rate of nucleation (*i.e.*, the number of new nodules appearing in unit time) per unit volume is constant; and (iii) as the reaction progresses, the cementite* dissolves and maintains a constant carbon concentration in the austenite, and the diffusion distance L remains constant. The condition visualized in assumption (iii) is shown diagrammatically in Fig. 7.

Considering a constant area of advancing interface and the radius r of a graphite sphere increasing by dr in time dt,

$$C_p \frac{\partial r}{\partial t} = C_e \frac{\partial r}{\partial t} + J \quad \ldots\ldots\ldots\ldots\ldots\ldots (11)$$

where J is the flux of carbon atoms down the concentration gradient. To simplify the mathematical

* The model does not require any assumptions to be made about the composition of the cementite.

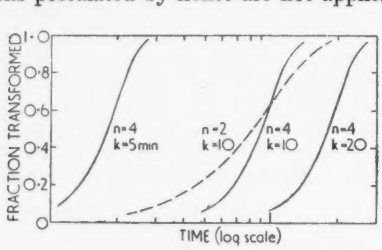


Fig. 5—Hypothetical reaction curves for various values of *n* and *K*

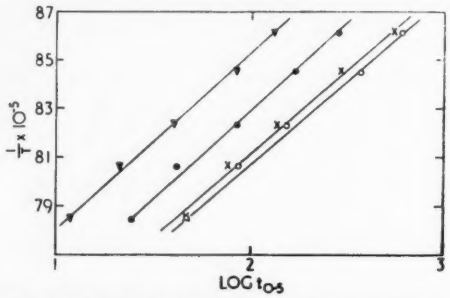
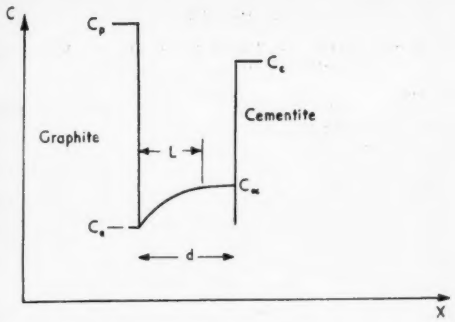


Fig. 6—Graph of log $t_{0\cdot5}$ *vs.* $1/T$ (for melt 35)



C = Carbon concentration
C_p = Composition of graphite
C_e = Composition of austenite in equilibrium with graphite
C_a = Composition of austenite in equilibrium with cementite
C_c = Composition of cementite

Fig. 7—Model used in kinetic analysis

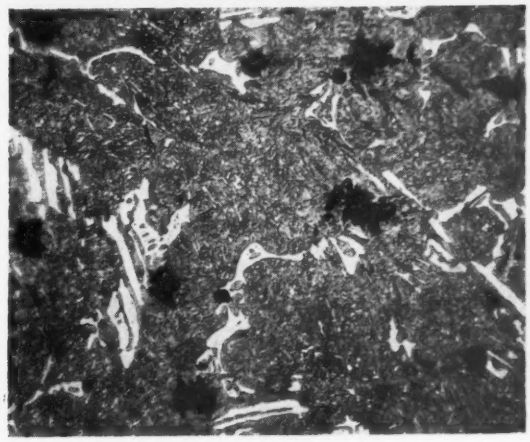


Fig. 8—Alloy containing 2·75% of carbon and 1·07 of silicon annealed for 2 hr. at 850° C. × 205

treatment the concentration gradient is assumed to be linear and the diffusion coefficient D is assumed to be independent of composition. Since

$$J = D\frac{\partial c}{\partial x} \text{ and } \frac{\partial c}{\partial x} = \frac{C_a - C_e}{L},$$

equation (11) becomes

$$\frac{\partial r}{\partial t} = \frac{D}{L}\left(\frac{C_a - C_e}{C_p - C_e}\right) = G \quad \ldots\ldots(12)$$

which is constant.

For a nodule nucleated at time t_1, the volume at time t will be $\frac{4}{3}\pi G^3(t - t_1)^3$ and the increase in volume after time dt will be

$$dv = 4\pi G^3(t - t_1)^2 dt \quad \ldots\ldots(13)$$

Since the number of nuclei formed from time t_1 to $(t_1 + dt_1)$ is $dn = N.dt_1$ per unit volume of material, the total increase in volume is

$$dV = 4\pi G^3(t - t_1)^2 N.dt_1.dt \quad \ldots\ldots(14)$$

However, the probability of completely free growth decreases as the transformation proceeds. It may be assumed that this probability is expressed by $c'(1 - y_t)$, where c′ is a constant.

The increase in volume of the graphite due to all the nuclei formed between times 0 and t is

$$dy_t = 4\pi G^3 N dt(1 - y_t)c'\int_0^t (t - t_1)^2 dt_1$$
$$= \frac{4}{3}\pi G^3 N c'(1 - y_t)t^3 dt \quad \ldots\ldots(15)$$

and hence

$$y = 1 - \exp(-c'\frac{\pi}{3}G^3 N t^4) \quad \ldots\ldots(16)$$

Table V

VALUES OF Q_K

Ingot No.	Q_K, cal./mole
19	58,000
27	74,000
33	74,000
35/1	67,000
35/2	67,000
35/3	64,000
35/4	64,000

and from equation (12)

$$y = 1 - \exp\left[-c'\frac{\pi}{3}\frac{D^3}{L^3}\left(\frac{C_a - C_e}{C_p - C_e}\right)^3 N t^4\right] \quad \ldots\ldots(17)$$

This may be written

$$y = 1 - \exp\left(-\frac{t}{K}\right)^n,$$

which is equation (1) where

$$n = 4 \text{ and } \frac{1}{K^n} = \left[c'\frac{\pi}{3}\frac{D^3}{L^3}\left(\frac{C_a - C_e}{C_p - C_e}\right)^3 N\right].$$

Thus, the value of n required by the model is almost identical with that obtained experimentally for first-stage graphitization.

Further experiments were carried out to examine each of the factors about which major assumptions have been made in the above analysis.

GENERAL MICROSCOPICAL EXAMINATION

Numerous specimens transformed for different times at different temperatures were examined microscopically. In every case it was found that the graphite was present as approximately spherical nodules. However, they were often not as perfect in form as those usually found in comparable commercial alloys. Quenching experiments showed that almost all the nodules were situated at, or very close to, an austenite–cementite interface. The small number that were situated in the austenite matrix (martensite at room temperature) could be attributed to the probability of the plane of polish intersecting some nodules that had been nucleated by carbide particles lying either just above or below the plane. Each graphite spherulite was surrounded by martensite, indicating that at temperatures above A_1 the graphite grows in a matrix of austenite through which the carbon must diffuse. The microstructure of a specimen annealed to an early stage in the reaction, which demonstrates these features, is shown in Fig. 8.

Thus, in general terms, the model is a suitable representation of the arrangement of phases during the reaction.

No attempt was made to measure the diffusion distance L, and so the validity of this assumption

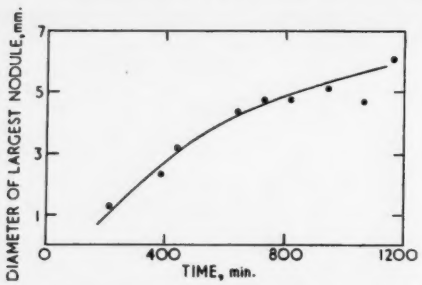


Fig. 9—Growth curve of melt 33 at 853° C.

can only be decided from indirect evidence. Provided that the dissolving cementite maintains a constant value of C_e (*see* Fig. 7), L may be expected to remain sensibly constant over a range of distance d, between the interfaces. Since the microscopic examination showed that the graphite nodules appear and grow in close proximity to the cementite, it seems that the assumption of a constant L is reasonable. At a late stage in the reaction, cementite particles remote from graphite start to dissolve before those adjacent to it have completely disappeared. This suggests that when this stage is reached, the neighbouring cementite is incapable of maintaining saturation at the periphery of the austenite sphere surrounding the growing graphite nodule, and the more remote cementite attempts to maintain the balance.

EXPERIMENTAL DETERMINATIONS OF RATES OF GROWTH AND NUCLEATION

The part of the ingot which remained after the dilatometer specimens had been removed was cut into small specimens suitable for heat-treatment and microscopic examination. To ensure uniformity in the conditions of vacuum and rate of heating, each specimen was heat-treated in the dilatometer tube.

The linear rate of radial growth G at a particular temperature was determined by measuring the diameter of the largest nodule found in each of a series of samples reacted for increasing time intervals. Plotting this diameter as a function of time gives a curve with a gradient equal to twice the radial rate of growth. The accuracy of the method depends upon examining samples with a sufficiently large number of nodules, so that the diameter of the largest visible circular patch of graphite may be assumed to be the major diameter of a nodule formed at the first instant of reaction. Each specimen was measured at a magnification of 168 dia. on at least five different planes obtained by repeated preparation. The results showed some scatter and were averaged.

A typical plot of the diameter of the longest visible nodule as a function of time is shown in Fig. 9. Attempts were made to represent the data by drawing a straight line through the points by the least-square method. However, in every case such a plot intersects the nodule-diameter axis at some positive value, implying that nodules of appreciable dimensions exist before the reaction starts. This is not feasible; in fact, the time axis should be cut at a positive value (the time of the incubation period). Unfortunately, owing to the difficulty of measuring very small

Table VI

VALUES OF THE 'AVERAGE' RADIAL RATE OF GROWTH, mm./min. × 168

Temperature, °C.	Ingot Number 33	35/1	35/2	35/3	35/4
811	0·00048				
853	0·0028				
888				0·009	0·012
900	0·010				
910		0·012	0·013		0·031
943		0·026	0·020	0·031	0·065
968	0·037	0·040	0·040	0·060	0·093
1004		0·065	0·070	0·100	

nodules, it is not possible to determine the early part of the growth curve. In general it is thought that the data are represented best by smooth curves of the type shown in Fig. 9. However, in no case is the deviation from linearity very large over the time interval in which experimental determinations were possible. The gradient of the best straight line through the experimental points may be considered as an 'average' value of the rate of growth and is of some value for purposes of comparison. The 'average' growth rates are listed in Table VI. These are increased both by raising the annealing temperature and by increasing the silicon content. The temperature/growth-rate relationship for ingots from melt 35 is shown in Fig. 10.

The rate of nucleation N is defined as the rate at which new nuclei appear. This may be obtained by determining the total number of graphite nodules in a series of specimens reacted for increasing periods of time. The gradient of the plot of the number of nodules per unit volume of material as a function of time gives the rate of nucleation per unit volume. This value can then be converted into terms of unit volume of untransformed matrix or unit austenite–cementite interfacial area.

The problem of deriving the number of precipitate particles per unit volume of sample from measurements made on a plane surface is a matter of some difficulty. Several approximate methods are available but these require certain assumptions that are unreasonable in the present case. The method that was used is based upon a mathematical analysis due to Scheil.[10] This requires the precipitate particles to be (i) spheres, (ii) randomly distributed, and (iii) of discrete sizes, of radius r_1, r_2, r_3, etc. Since none of these requirements is fulfilled exactly, it is not surprising that the results exhibit some scatter. The largest error is probably introduced by the assumption

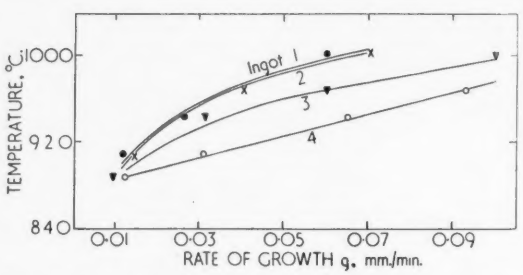


Fig. 10—Effect of temperature on rate of growth (melt 35)

Table VII

VALUES OF THE RATE OF NUCLEATION, no. of nodules/cu. mm./min.

Temperature, °C.	Ingot Number 33	35/1	35/2	35/3	35/4
811	46				
853	35				
888				114	
900	148				
910		16	29		247
943		89	73	147	288
968	713	140	260	617	
1004		200	540	766	

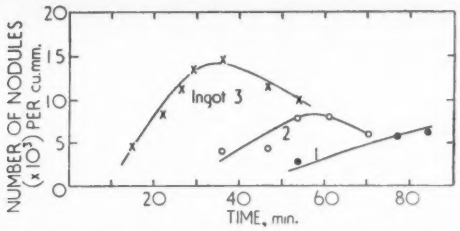


Fig. 11—Nucleation curves of melt 35 at 968° C.

of random distribution. Since nucleation occurs at the austenite–graphite interface, it would be more logical to express N in terms of the number of nodules per unit interfacial area. However, in practice it proved impossible to obtain even an approximate estimate of this area.

Nucleation curves were determined experimentally using the growth-measurement specimens. A total of 18 curves was determined. Representative curves are shown in Fig. 11. Eleven curves indicated that, with increase in time, the number of nodules first increases, reaches a maximum, and subsequently decreases. The remainder showed a continuously increasing number of nodules. These might pass over a maximum at a later stage of graphitization. In all cases the portion of the curve preceding the maximum is approximately linear. The existence of a maximum indicates that after a certain reaction time some graphite nodules (presumably the smallest) redissolve. This rather surprising result was noted by Brown and Hawkes[11] in a few nucleation curves which they determined for the graphitization of commercial white irons. They put forward the very reasonable explanation that the phenomenon was a consequence of the dependence of solution potential on the radius of the graphite spheres.

Silicon has a considerable effect on the nucleation. Figure 11 shows that increasing the silicon content increases the initial slope of the curve and moves the maximum to shorter time and greater number of nodules. Thus, when reacted for equal time-intervals, specimens with higher silicon contents contain a greater number of nodules.

In each case, the initial section of the curve was approximately linear and the slope can be regarded as the rate of nucleation. Values determined in this manner are tabulated in Table VII. Comparison of Tables VI and VII shows that silicon has a much greater effect on the rate of nucleation than on the rate of growth. For example, at 910° C. the rate of growth in alloy 35/4 (1·37% of Si) is $2\frac{1}{2}$ times as great as that in alloy 35/1 (0·86% of Si), whereas the rate of nucleation differs by a factor of 15. It is also evident that the influence of temperature on the rate of nucleation is greater than its effect on the rate of growth.

GENERAL DISCUSSION

In Zener's[2] analysis of the kinetics of precipitation from a uniform solid solution of infinite extent, L varies continuously with time, and a parabolic time-dependence of G is predicted. The model for the graphitization reaction differs basically from the Zener analysis in that constant L and G with time are postulated. The constancy of L can only be inferred, but the validity of the supposition is strongly supported by the experimental demonstration that the growth curves deviate only slightly from a straight line. Thus, it seems that no serious discrepancy between the ideal index n and that determined experimentally is attributable to these factors. However, the other assumptions on which the model relies have been found to rest on less convincing experimental evidence. The nucleation curve is approximately linear only up to a time approaching the position of the maximum. The assumption of a spherical shape for the nodules is not completely supported by observation, many having a somewhat irregular shape. This fact would also tend to increase the discrepancy between the experimental and ideal values of n. Finally, the assumption that the probability of 'soft impingement' (and hence the decrease in the rate of growth) is given by the factor $c'(1 - y)$ is an over-simplification. The correct probability factor is complex and depends upon a number of interrelated factors; it will be different in almost every specimen. Thus, it appears that the remarkably close agreement between the predicted and experimental values of n is in part fortuitous, resulting from the existence of compensating errors. However, it is considered that the experimental results amply justify the contention that the model which has been suggested is a reasonable, although somewhat idealized, representation of the mechanism of the graphitization reaction.

ACTIVATION ENERGY FOR NUCLEATION

As a first approximation the rate of nucleation may be expressed in the form[12]

$$N = c'' \exp(-Q_D/kT).\exp(-A'/kT) \ldots\ldots (18)$$

where Q_D is an activation energy for diffusion and is independent of temperature and A' is related to the free energy (ΔF_C) required to form a nucleus of the precipitating phase and is generally a function of temperature. For a spherical nucleus, ΔF_C is proportional to the cube of the surface energy and is inversely proportional to the square of the volume free-energy. It is reasonable to assume that the surface energy is approximately independent of temperature. In most precipitation processes the volume free-energy is increased as the reaction temperature is lowered, so that $A' \sim 1/(T_E - T)^2$, where T_E = equilibrium temperature and T = reaction temperature. However, in the case of first-stage graphitization, $(T_E - T)$ is sensibly independent of the value of T.

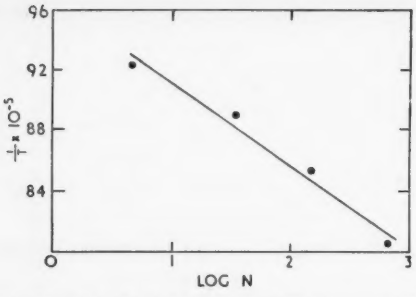


Fig. 12—Graph of log N *vs.* $1/T$ (for melt 33)

This may be appreciated by considering the relative position of the stable and metastable austenite solubility lines in the well-known double iron–carbon diagram.[13] In this special case, $(T_E - T)$, and hence A', may be considered to be a constant independent of temperature. Thus equation (18) becomes

$$N = c''.\exp[-(Q_D + A')/kT],$$

predicting that the rate of nucleation increases exponentially with temperature. In Fig. 12 representative experimental results are plotted in the form $\log N$ *vs.* $1/T$. All such plots showed only small deviations from linearity, suggesting that, as a first approximation, equation (18) is applicable. From the slope of the plot the value of $(Q_D + A')$ may be calculated: this proves to be 79,000 cal./mole. Assuming Q_D to be 26,000 cal./mole (activation energy for carbon diffusion in austenite), the activation energy to form a graphite nucleus A' is 53,000 cal. By considering A' to be constant in equation (18) it follows that the isothermal transformation curves will not exhibit the usual 'C' curve behaviour, but rather that the reaction time will decrease continuously as the temperature is increased. The experimental curves (*see* Fig. 2) support this deduction.

RATE-CONTROLLING FACTOR

Application of the Boltzmann distribution law to the statistical theory of rate processes in the solid state leads to the general formula

$$v = A_1 \exp\left(-\frac{Q}{RT}\right) \quad \ldots\ldots\ldots\ldots\ldots\ldots (19)$$

where v is the rate of reaction, Q is the activation energy, A_1 is a constant, and the other symbols have the usual meaning. The value of Q may be calculated from suitable experimental data by plotting $\log v$ against $1/T$ and measuring the slope, which equals $-Q/2{\cdot}3R$. It is unusual for the accuracy to be better than $\pm$ 6000 cal./mole. The following average activation-energy values were determined from the experimental data:

Graphitization reaction	Q_K = 68,000 cal./mole
Nucleation only	Q_N = 79,000 cal./mole
Growth only	Q_G = 63,000 cal./mole

In reactions that are entirely growth-controlled the activation energy for the complete reaction will be identical with that for the controlling diffusion process. In the more general case of a nucleation and growth reaction this relation does not hold, since the total activation energy depends upon both these functions. It has been shown that first-stage graphitization conforms to the semi-empirical equation

$$y = 1 - \exp\left(-\frac{t}{K}\right)^n \quad \ldots\ldots\ldots\ldots\ldots\ldots (1)$$

The value of Q_K is obtained from the temperature-dependence of K. The model indicates that the reaction rate depends upon a constant rate of nucleation and the cube of a constant rate of growth. Explicitly,

$$\frac{dy}{dt} = A_2(1 - y)t^3 \quad \ldots\ldots\ldots\ldots\ldots\ldots (15)$$

where

$$A_2 = \frac{4}{3}\pi G^3 N c' \quad \ldots\ldots\ldots\ldots\ldots\ldots (20)$$

This predicts that

$$Q_{A_2} = 3Q_G + Q_N \quad \ldots\ldots\ldots\ldots\ldots\ldots (21)$$

From equations (2) and (15),

$$A_2 = n/k^n \quad \ldots\ldots\ldots\ldots\ldots\ldots (22)$$

$$Q_{A_2} = nQ_K \quad \ldots\ldots\ldots\ldots\ldots\ldots (23)$$

Thus, the experimental data give a value for Q_{A_2} of 272,000 cal./mole, which correlates closely with the experimental value for $(3Q_G + Q_N)$ of 268,000 cal./mole. This provides further evidence of the suitability of the model.

Since Q_K is a compound function, comparison with known activation energies cannot establish the overall rate-controlling process. However, it is profitable to consider the growth process separately. Growth of graphite in high-purity iron–carbon–silicon alloys involves three diffusion processes. Carbon diffuses from the cementite, through the austenite, to the graphite interface. Since the graphite is pure carbon,[14] silicon will build up in front of the advancing interface. As the activity of carbon is increased by the presence of silicon, this accumulation will reduce the carbon activity gradient and will consequently retard the diffusion of carbon. For the reaction to proceed, silicon must diffuse away from the graphite interface. In addition, iron atoms must diffuse away from certain locations in the austenite to provide the necessary free volume for the growing graphite.

From equation (12) it will be seen that G, the rate of growth, is proportional to D, the diffusion coefficient, and hence Q_G should be equal to Q_D, where Q_D is the activation energy of the controlling diffusion process. The study of carbon diffusion in gamma iron by Wells, Batz, and Mehl[15] showed that Q_C decreases as the carbon concentration is increased, the value for saturated austenite being about 26,000 cal./mole. Thus, it is immediately obvious that carbon migration cannot be the controlling process. Furthermore, these authors showed that up to 2% of silicon had little effect on the carbon diffusion coefficient (and consequently on Q_C), and so the high value of the activation energy for graphitization cannot be explained in terms of the effect of silicon on Q_C for carbon diffusion.

There are few reliable data about the other two diffusion processes. Smithells[16] gives two values for the diffusion coefficient of silicon in gamma iron, from which it can be calculated that Q_{Si} is 56,000 cal./mole. However, Batz, Mead, and Birchenall[17] state that the values of D given by Smithells are too

high, and they determine values from which it can be calculated that Q_{Si} is 77,000 cal./mole. A value for Q_{Fe}, the self-diffusion of iron, has been reported by Birchenall and Mehl[18] as 74,000 cal./mole, but it is very doubtful if this can be considered as valid under conditions other than those existing in pure iron. Thus it seems that the activation energy of the growth process is of the same order of magnitude as that for diffusion of silicon and iron. The controlling process cannot be established unequivocally until more accurate diffusion data are available.

SUMMARY

The results obtained and the conclusions drawn from them can be summarized as follows:

(1) First-stage graphitization takes place by a process of nucleation and growth, the nuclei appearing at the interface between cementite and austenite. Growth takes place by diffusion of carbon through austenite.

(2) The incubation period and the total time for first-stage graphitization decrease logarithmically with rise of temperature. The plots are parallel.

(3) A linear relationship exists between the logarithm of the graphitization time and the logarithm of the silicon content.

(4) Isothermal graphitization conforms to the equation

$$y = 1 - \exp\left(-\frac{t}{K}\right)^n.$$

(5) The value of n is approximately 4 and is independent of temperature and composition.

(6) The isothermal curves obtained over wide ranges of temperature and composition can be superimposed by moving the zero of the time scale.

(7) The constant K has an exponential relation with temperature, and the activation energy for graphitization Q_K is 68,000 cal./mole.

(8) The rate of growth of graphite decreases continuously with time.

(9) Increasing the silicon content increases slightly the average rate of growth.

(10) The rate of growth increases as the temperature rises.

(11) When the number of graphite nodules is plotted as a function of time, the graph shows a maximum.

(12) Silicon markedly increases the rate of nucleation.

(13) As the silicon content is increased, the nodule number increases and the position of the maximum moves to a shorter time and a greater number of nodules.

(14) The rate of nucleation increases with rise in temperature.

(15) The influence of temperature on the rate of nucleation is greater than its influence on the rate of growth.

(16) The relative effects of silicon content on the rates of growth and nucleation are similar to those of temperature.

(17) The rate of growth and the rate of nucleation have exponential relations with temperature (approximately). The activation energy for growth is 63,000 cal./mole and that for nucleation is 79,000 cal./mole.

(18) The observed kinetic behaviour can be accounted for in terms of a model which assumes spherical nodules, a constant diffusion distance, and constant rates of nucleation and growth. The relationship between the activation energies for graphitization, for growth, and for nucleation predicted by the model is closely obeyed.

(19) It is reasonable to suppose that the rate of growth is controlled by the rate of silicon diffusion, although owing to uncertainty in the published diffusion data the possibility that diffusion of iron is the vital process cannot be discounted.

Acknowledgments

The work described in this paper was carried out in the Department of Metallurgy, University of Liverpool. The authors wish to thank the Andrew Carnegie Research Fund and the Department of Scientific and Industrial Research for financial assistance. One of the authors (J.B.) held a Munitions Fellowship (University of Liverpool). Thanks are also extended to the British Cast Iron Research Association for analysis of all the alloys used in the investigation.

References

1. J. Burke and W. S. Owen: *J. Iron Steel Inst.* (to be published).
2. C. Zener: *J. Appl. Phys.*, 1949, vol. 20, pp. 950–953.
3. C. S. Roberts, B. L. Averbach, and M. Cohen: *Trans. Amer. Soc. Metals*, 1953, vol. 45, pp. 576–599.
4. R. W. Balluffi, M. Cohen, and B. L. Averbach: *Ibid.*, 1951, vol. 43, pp. 497–517.
5. B. L. Averbach, M. Cohen, and S. G. Fletcher: *Ibid.*, 1948, vol. 40, pp. 728–754.
6. C. A. Wert: *J. Appl. Phys.*, 1949, vol. 20, pp. 943–949.
7. W. A. Johnson and R. F. Mehl: *Trans. Amer. Inst. Min. Met. Eng.*, 1939, vol. 135, pp. 416–442.
8. J. H. Hollomon: Private communication.
9. C. Zener: *Trans. Amer. Soc. Metals*, 1949, vol. 41, pp. 1057–1058.
10. E. Scheil: *Zeitschrift für anorganische und allgemeine Chemie*, 1931, vol. 201, pp. 259–264.
11. B. F. Brown and M. F. Hawkes: *Trans. Amer. Found. Soc.*, 1951, vol. 59, pp. 181–200.
12. R. Becker: *Proc. Phys. Soc.*, 1940, vol. 52, pp. 71–76.
13. L. S. Darken and R. W. Gurry: *Trans. Amer. Inst. Min. Met. Eng.*, 1951, vol. 191, pp. 1015–1018; *J. Met.*, 1951, vol. 3, Nov.
14. W. S. Owen and B. G. Street: *J. Iron Steel Inst.*, 1951, vol. 167, pp. 113–116.
15. C. Wells, W. Batz, and R. F. Mehl: *Trans. Amer. Inst. Min. Met. Eng.*, 1950, vol. 188, pp. 553–560; *J. Met.*, 1950, vol. 2, Mar.
16. C. J. Smithells: "Metals Reference Book": 1949, London, Butterworths Scientific Publications.
17. W. Batz, H. W. Mead, and C. E. Birchenall: *Trans. Amer. Inst. Min. Met. Eng.*, 1952, vol. 194, p. 1070; *J. Met.*, 1952, vol. 4, Oct.
18. C. E. Birchenall and R. F. Mehl: *Trans. Amer. Inst. Min. Met. Eng.*, 1950, vol. 188, pp. 144–149; *J. Met.*, 1950, vol. 2, Jan.

Determination of Tin

By the Methods of Analysis Committee

SYNOPSIS

Omission of the sulphide separation in British Standard 1121: Part 20: 1951 provides a simplification by which results similar to those given by Part 7 can be obtained. High copper and phosphorus contents have been shown to be without effect on this procedure. 856

BRITISH Standard methods for the determination of tin in iron and steel have been published as B.S.1121: Part 7: 1948 "Tin in pig iron, plain carbon steels and certain low-alloy steels" and B.S.1121: Part 20: 1951 "Tin in highly alloyed steels." It was felt that the provisions of Parts 7 and 20 were not mutually exclusive, the main differences in Part 20 being the solvent used, which included sulphurous acid, and the introduction of a sulphide-separation stage. The Sub-Committee (whose constitution is given in the Appendix) originating the proposals in Part 7 was asked to investigate the possibility of combining these procedures into a single comprehensive method.

Paper MG/D/65/53 of the Methods of Analysis Committee of the Metallurgy Division of B.I.S.R.A., received 15th July, 1953.

Table I

RESULTS BY FIRST MODIFICATION OF B.S.1121: PART 20: 1951

MGS21 Mild steel, 0·05% of molybdenum MGS38 Mild steel, 0·2% of copper MGS198A 3% nickel–chromium steel

Analyst	Tin, %			Remarks
	MGS21	MGS38	MGS198A	
1	0·031	0·032	0·18	No sulphurous acid
	0·033	0·033	0·17	Residual " "
	0·030	0·031	0·16	5 ml. of excess sulphurous acid
	Nil	Nil	Nil	25 ml. of " " "
2	0·031	0·031	0·17	No sulphurous acid
	0·030	0·032	0·17	Sulphurous acid boiled off
	0·032	0·030	0·17	As above, oxidized with bromine
3	0·031	0·030	0·17	Excess sulphurous acid boiled off
	0·030	0·031	0·17	
4	0·030	0·032	0·18	Residual sulphurous acid
	0·031	0·030	0·18	As above, oxidized with bromine
	0·029	0·030	0·18	As above, oxidized with hydrogen peroxide
5	0·027	0·027	0·16	Residual sulphurous acid
	0·028	0·026	0·16	
6	0·031	0·030	0·16	As above
	0·031	0·030	0·16	
7	0·027	0·028	0·17	As above
	0·027	0·028	0·17	
	0·028	0·029	...	
8	0·030	0·031	0·18	No sulphurous acid
	0·030	0·031	0·18	Residual sulphurous acid
	0·030	0·031	0·18	As above, oxidized with bromine
9	...	...	0·17	
10	0·029	0·032	0·18	No sulphurous acid
	0·029	0·029	0·17	Residual sulphurous acid
	0·026	0·029	0·17	Oxidized with bromine
11	0·026	0·025	0·18	No sulphurous acid
	0·026	0·024	0·17	Residual " "
	0·026	0·024	0·17	As above, oxidized with bromine
Average using Part 7	0·030	0·028	0·175	

Table II

EFFECT OF COPPER

Analyst	Tin, % (Sample MGS198A) As-received	0·5% of Copper Added
1	0·17	0·175
2	0·175	0·18
3	0·17	0·175
4	0·18	0·18
6	0·165	0·165
7	0·165	0·17
8	0·175	0·175
9	0·17	0·17

Table III

EFFECT OF PHOSPHORUS

Analyst	Tin, % (Sample MGS199) As Method	5 ml. of Bromine Added
1	0·34	0·34
2	0·32	0·31
3	0·31	0·31
4	0·31	0·34
8	0·31	0·31
9	...	0·31
10	0·30	0·31
11	0·31	0·31

At the same time, reports had been received of discrepancies in the behaviour of certain types of phosphoric iron when analysed under these conditions, and confirmation of such claims was sought.

PRELIMINARY WORK

An examination of Part 20 indicated that the omission of hydrogen sulphide would provide a method similar in principle to that in Part 7, and with this modification Part 20 was used to determine tin on two samples of mild steel, MGS21 and MGS38, the latter containing 0·2% of copper, and on a sample of alloy steel MGS198*A*. Table I shows the results obtained, which agree very closely with analyses previously made using Part 7.

It was suggested that the presence of sulphurous acid might interfere with the determination, and in some tests recorded in Table I sulphur dioxide was removed by boiling or by oxidation. Trials with a large excess of the reagent further confirmed the need for its removal.

INTERFERENCE OF COPPER

To investigate a suggestion that copper was being reduced by the aluminium/antimony mixture, thus influencing the operation of the method, further trials were made on sample MGS198*A* with the addition of 0·5% of copper. The absence of significant effect is shown by the results given in Table II.

INTERFERENCE OF PHOSPHORUS

In view of reports that difficulty had been met with in obtaining consistent figures by Part 7 on melting samples of high-phosphorus iron, a sample of this material (1·4% of phosphorus) was circulated and the effect of a proposed modification to the method (the addition of bromine to the solution before reduction with aluminium) was examined. This precaution was suggested because of a suspicion that phosphorus compounds present in the reduced state were affecting the final titration. The results obtained, given in Table III, did not, however, indicate that any greater reproducibility could be achieved by this addition.

CONCLUSIONS

The provisions of B.S.1121: Part 20: 1951* can be satisfactorily applied to the determination of tin on the whole range of iron and steel compositions. For iron and carbon and low-alloy steels, sulphide separation of the tin is unnecessary, but if direct reduction and titration is used the addition of sulphurous acid must be omitted or the excess must be completely removed by boiling. There is no evidence of interference by copper or phosphorus.

APPENDIX

Constitution of the Sub-Committee at 31st December, 1952

S. W. CRAVEN (*Chairman*)	I.C.I. Ltd. (Alkali Division)
O. BEADLE	Cargo Fleet Iron Co., Ltd.
W. E. CLARKE	British Cast Iron Research Association
C. CLYDE	Lanarkshire Steel Co., Ltd.
L. E. GARDNER	Edgar Allen and Co., Ltd.
C. L. GRAYSON	English Steel Corp., Ltd.
R. F. JONES	Steel Company of Wales Ltd.
A. W. REED	Consett Iron Co., Ltd.
N. D. RIDSDALE	Bureau of Analysed Samples Ltd.
C. W. SHORT	John Lysaght Ltd.
F. SPEED	Dorman, Long and Co., Ltd.
H. J. TURNER	Steel, Peech and Tozer

* *See J. Iron Steel Inst.*, 1950, vol. 165, pp. 190–197.

Knurling of Cogging Rolls

By A. H. Norris

In common with many other works, it has been the practice at the Normanby Park Steelworks of John Lysaght Ltd. to 'rag' the cogging rolls in the initial passes. The passes were grooved with either a chipping hammer or a deseaming torch.

No ragging was carried out on the rolls for a long period, but recently the whole matter had to be reconsidered, for two reasons: firstly, owing to the large quantity of free-cutting steels rolled, excessive slipping was taking place, reducing outputs by as much as 30% in extreme cases, and secondly, owing to slippage, the rolls were 'shuddering,' with consequent

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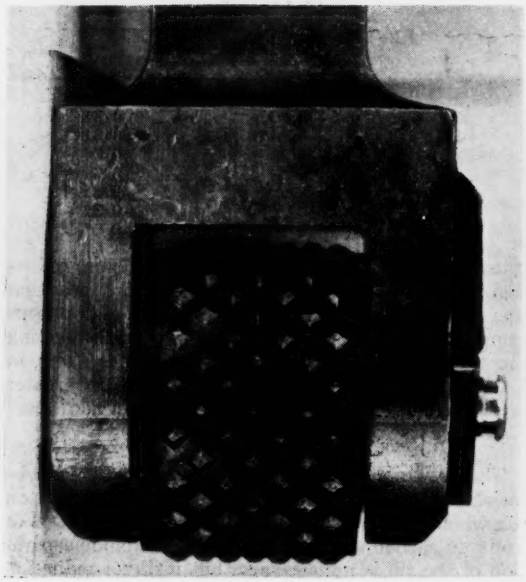


Fig. 1—Knurling tool mounted in holder

detriment to bearings, spindles, and mechanical gear generally.

In collaboration with the supplier, experiments were carried out with a special knurling tool, shown in Fig. 1. The tool itself is supported in front and behind on props and is wedged under the tool bridge, pressure being applied from the rear as the tool cuts into the roll. The revolving action of the roll makes the knurling wheel rotate, marking the roll as shown in Fig. 2.

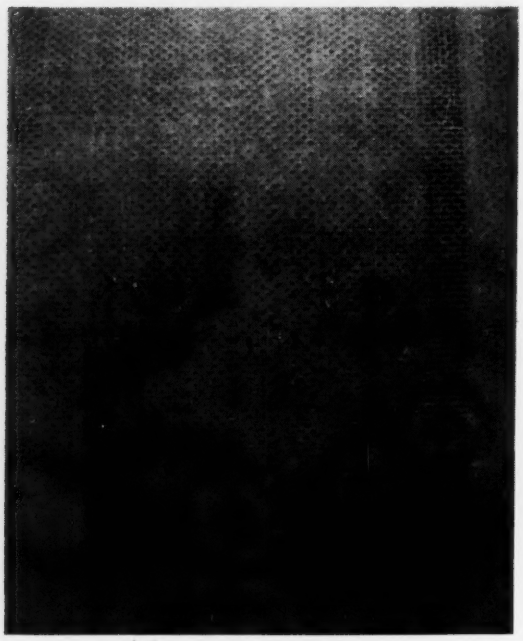


Fig. 2—Top blooming-mill roll after dressing, showing knurling. This roll has rolled 95,196 tons, and is $1\frac{7}{8}$ in. dia. below original size

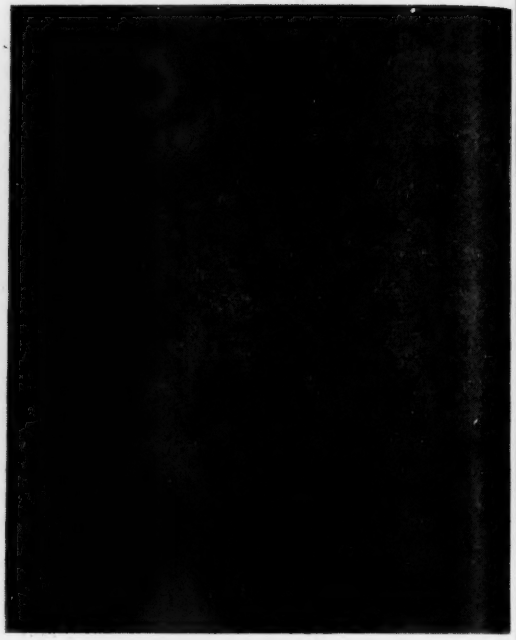


Fig. 3—Top blooming-mill roll, after 3 weeks' rolling (30,455 tons); total tonnage rolled 207,899 tons, $3\frac{5}{8}$ in. dia. below original size

The knurling tool itself consists of a mild-steel holder, a pin made of case-hardened steel on which the wheel revolves, and the knurling wheel, which is of high-speed tool steel with small pyramidal protrusions at regular intervals over the surface. The wheel in present use is 2 in. wide × $2\frac{3}{4}$ in. dia.; the protrusions on its surface are $\frac{3}{16}$ in. square at the base, tapering to a point $\frac{1}{8}$ in high, the distance between centres being $\frac{5}{16}$ in.

Since February, 1953, the cogging rolls for the first three passes have been knurled in this way with considerable success. Outputs on the initial shifts of the week after newly dressed rolls have been mounted have been affected only slightly by slippage, and 'jarring' to the rolls has been greatly reduced. No sparks or pieces of scale fly from the rolls, as with ragged rolls, and—most important—there has been less fire-cracking on the passes concerned. It is believed that by knurling the passes fire-cracking has been reduced, even in comparison with rolls that have been mounted plain without ragging or knurling. Figure 3 shows the ingot pass in a roll taken out for dressing after rolling over 30,000 tons since the previous dressing, and with a roll life of 208,000 tons. The fire-cracking is more intermittent than before and less severe.

Experiments are still proceeding, but it is believed that the introduction of knurling will increase roll life and reduce fire-cracking, with small loss of output due to slipping.

Some Factors Affecting the Wear of Graphite Electrodes in the Electric-Arc Furnace

By the Electric Process Sub-Committee

SYNOPSIS

A co-operative investigation by members of the Electric Process Sub-Committee and the research staffs at their plants has shown that abnormally rapid wear of graphite-electrode collars is usually associated with a high electrical resistivity, which is probably due to incomplete graphitization of the material. The heating of joints is likely to be aggravated by certain features of nipple and socket construction when the resistivity is high.

Simple non-destructive resistivity tests for complete electrodes have been devised and are described. These tests indicate that in the case of a ' bad ' electrode the resistivities of the ends are often higher than that of the middle. 854

In comparing electric-arc furnaces with fuel-burning furnaces (such as the open-hearth) for the manufacture of alloy steels, electric power and electrode consumption can be regarded as fuel costs. These costs are one of the chief obstacles in the way of the expansion of electric steelmaking in many countries. In Great Britain, typical power costs using a two-slag process may vary around 50*s*. and electrode costs around 25*s*. per ton of steel made, according to conditions;* these costs compare very unfavourably with a typical open-hearth fuel cost of about 30*s*. per ton for alloy steels. The Electric Process Sub-Committee of the British Iron and Steel Research Association has as one of its main objectives the reduction in the cost of steelmaking in arc furnaces, and has on hand a number of projects designed to effect economies in the consumption of electrodes. The programme includes economizer design, the influence of current density on rate of wear, the use of amorphous carbon and of Söderberg electrodes as alternatives to graphite, the use of coated electrodes to reduce oxidation, and the effect of changes in the chemical and physical properties of graphite electrodes on their rate of wear.

The consumption of electrodes in arc furnaces depends on a large number of factors, including:

(i) The time taken by the heat, which itself depends on many factors, such as the rate of power input and the quality of the steel being made

(ii) The quality of the raw materials, *e.g.*, the extent to which oxidized scrap is used

(iii) Furnace-design factors, such as the roof height and electrode diameter (which govern the surface area of electrode exposed within the furnace), the design of the economizers, and the response of the controllers

(iv) The care with which electrodes are handled and furnace parts are maintained

(v) The quality of the electrode material itself and the accuracy of the machining at the joints and nipples.

Some 44 years ago, Hering[1] drew attention to the effect of electrode characteristics so far as economy in heat losses was concerned, and recently Pensa[2] has discussed all aspects of electrode usage. The present paper deals mainly with item (v), which has proved to be an important factor affecting consumption. As part of its research programme the Committee began works trials designed to ascertain the effect of many of these factors on electrode consumption. Since the variations that can be imposed on everyday practice are limited, correspondingly small differences in performance have to be detected, and it was found that in many cases the effect of the variable under study was swamped by variations in electrode quality. Further studies were made in which members undertook new investigations and furnished information on earlier independent work; the results of these are given in this paper.

BEHAVIOUR OF GRAPHITE ELECTRODES

Since conditions vary so widely, appreciable variation in electrode consumption figures is to be expected. Nevertheless, Fig. 1 shows that large variations can occur on individual furnaces, and an investigation showed that these can be partly accounted for by changes in the quality of the electrodes.

When electrode consumption is abnormally high for the particular steelmaking conditions obtaining, the trouble most frequently experienced shows first as a reduction in the diameter of an electrode near a joint, usually known as ' necking.' The resulting weakening of the joint is liable to cause a premature breakage if the operator has not already anticipated this by removing the lower length of electrode. In either case, if it occurs frequently, the time taken to fit replacements can mean a serious loss of production time, which is perhaps even more costly than the value of the additional electrodes used.

Figure 2 is an example of an electrode that has become ' necked '; the collar, normally 3 in. thick, has been reduced in some places to as little as 1 in. Although this necking generally occurs below roof level, it is sometimes observed above the roof, the joint showing obvious overheating. Less commonly, a whole electrode length shows abnormal reduction of diameter, as in Fig. 3, although even in this case

* A large number of figures for power and electrode consumptions has been quoted by the Sub-Committee in a previous paper (*J. Iron Steel Inst.*, 1949, vol. 162, pp. 57–78).

Paper SM/AC/106/52 (Revised) of the Electric Process Sub-Committee of the Steelmaking Division of the British Iron and Steel Research Association, received 21st August, 1953.

Table I

ELECTRICAL RESISTIVITY RESULTS ON SAMPLES FROM BATCHES OF ELECTRODES DURING A PERIOD OF HIGH CONSUMPTION

Electrode	Electrical Resistivity, microhms/c.c. at 20° C. Longitudinal	Transverse	Mean
A	2100	2540	2130
B	1750	1450	1600
C	1300	2060	1680
D	1200	1350	1275

the reduction is more pronounced at the collars. The diameter of the lower length at the middle is about 12½ in. compared with the original 16 in.

TESTING OF ELECTRODES

The results of a comprehensive examination of electrode quality undertaken in 1942 when comparing certain specimens of British and imported electrodes were made available to the Sub-Committee.* The tests applied were as follows:

(i) Mechanical strength (*a*) Tensile
(*b*) Compression
(*c*) Torsional shear
(ii) Density (*a*) Bulk
(*b*) True
(iii) Porosity (*a*) Total
(*b*) Open
(*c*) Closed
(iv) Electrical resistivity
(v) Thermal conductivity
(vi) Specific heat
(vii) Thermal expansion
(viii) Spalling resistance
(ix) Burning rate
(x) Arc test
(xi) X-ray examination for constitution and crystal size
(xii) Ash content
(xiii) Moisture content.

Only tensile strength, electrical resistivity, and X-ray examination showed any relation with performance in service and, of the two former tests, the measurement of electrical resistivity is the simpler and can be applied to the non-destructive testing of a complete electrode. X-ray examination provided evidence of incomplete graphitization associated with high resistivity and poor performance. Spectrographic investigation showed that the ' bad ' electrodes compared favourably with ' good ' electrodes in regard to chemical purity, so that this factor does not seem to play a significant part.

Thus, the main conclusion from this work is that in testing it is sufficient to apply a resistivity test, and that X-ray examination for crystal structure might indicate the degree of graphitization.

Later, Heselwood and Bagnall examined specimens of electrodes that had shown evidence of abnormal wear at the collars. Both longitudinal and transverse resistivity tests were made and the results are given in Table I. These resistivity figures should be compared with those in Table III, which refer to later tests carried out on satisfactory electrodes. Results, including some for specimens from the middle of the electrode,† are shown in Table II.

Figures published by the manufacturers give the specific resistance of graphite as 760–1370 microhms/c.c.,‡ so that most of the specimens taken from electrodes showing abnormal wear exceeded these limits.

Since amorphous carbon has a considerably higher resistivity than graphite (about 3600–4600 microhms/c.c.), an X-ray examination of certain specimens was made to find out whether variations in electrical resistivity were associated with differences in the degree of graphitization. Figure 4 is an X-ray photograph of three specimens which gave resistivity figures of 1993, 1720, and 1200 microhms/c.c., respectively. The progressive improvement in the definition of the graphite lines (marked by arrows) indicates either a progressive increase in the proportion of the material that has been fully graphitized or greater graphitization of the material as a whole. This confirms that graphitization changes can be detected by resistivity measurements.

Although every specimen from the collars of electrodes showing abnormal wear in this region has given a high resistivity figure, the electrode in Fig. 3 showed reduction along its whole length, despite the fact that the high resistivity was confined to the ends; a specimen from the middle gave the normal figure of 1170

* By Mr. W. C. Heselwood (The United Steel Companies, Ltd.) and Mr. F. T. Bagnall (Samuel Fox and Co., Ltd.).

† Submitted by Mr. R. V. Thomas (Stewarts and Lloyds Ltd.).

‡ To convert to microhms/c.c. from microhms/cu. in., multiply by 2·54.

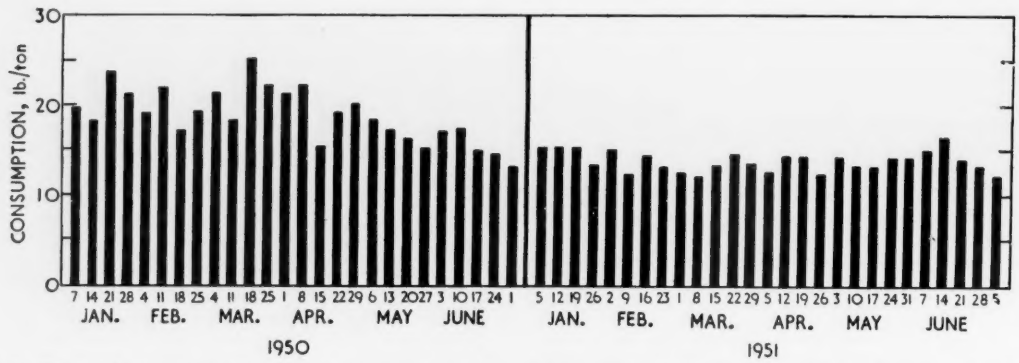


Fig. 1—Electrode consumption on a 12½-ton arc furnace (nominal capacity 10 tons) using 12-in. dia. graphite electrodes (Rotadyne control)

Fig. 2—Electrode collar after 'necking'

Fig. 3—Electrode showing abnormal reduction of diameter

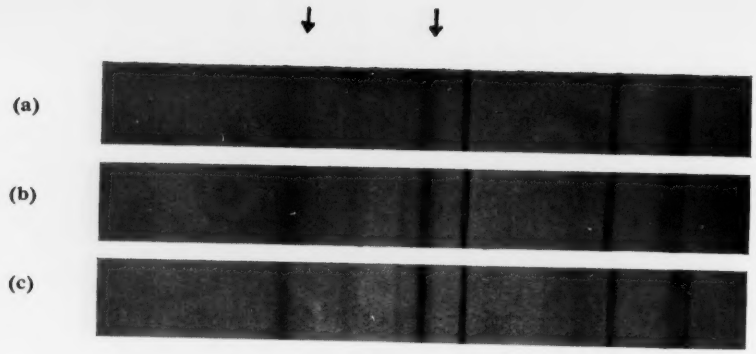


Fig. 4—X-ray photographs of electrode specimens: (*a*) Very poorly graphitized; (*b*) graphitization more marked than in (*a*); (*c*) most complete graphitization

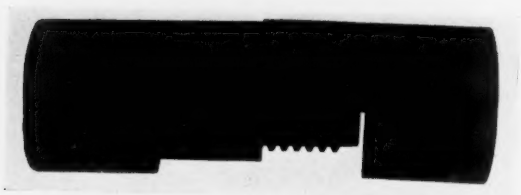


Fig. 5—'Cutaway' joint of 4-in. dia. electrode

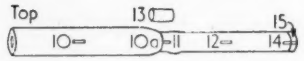
Fig. 6—General arrangement for measuring electrical resistivity of central portion of graphite electrodes

Table II

ELECTRICAL RESISTIVITY RESULTS FROM MIDDLE OF ELECTRODE

Spec. No.	Specific Resistance, microhms/c.c.
10	1120
10a	1720
11	1993
12	1233
13	1245
14	1523
15	1515
16*	2280
17*	2205

Key



* Specimens 16 and 17 were taken from the collars of another electrode that had shown abnormal wear

microhms/c.c. A high resistivity may be only one of the factors that can cause abnormally rapid wear, but it is probably the most common.

MECHANISM OF WEAR

It is not immediately obvious whether this rapid oxidation of incompletely graphitized specimens is due to differences in the rates of combustion of graphite and of, say, amorphous carbon, or whether it can be attributed to the higher temperatures resulting from the higher electrical resistance. There is evidence that amorphous carbon oxidizes more readily than graphite,[3] but the difference is not very great; since fairly small differences in degrees of graphitization are concerned it is unlikely that this can explain the large differences in rates of electrode wear sometimes encountered. Support for this view is provided by the results of burning-rate tests made independently by Heselwood and Thomas, where no correlation was found to exist between burning rate and electrical resistivity.

The combustion rates of all forms of carbon increase rapidly with temperatures up to 1100° C. Between 900° and 1100° C., for example, in a slow-moving air stream (3 cm./sec.), the combustion rate increases fivefold.[4] But calculations of approximate temperature rises likely to be encountered in practice on the assumption that the current is equally distributed over the cross-section of an electrode show that the differences due to variation in resistivity of the order found are not likely to be sufficient to account for the differences in rate of wear actually experienced.

The effect may possibly be explained as follows: nipples are of necessity slightly shorter than the sum of the lengths of two sockets if the ends of the electrodes are to make good electrical contact, so that there must be a gap at either or both ends. Figure 5 shows a ' cutaway ' joint of a 4-in. dia. electrode where the nipple has been screwed hard against the bottom of one socket, leaving a gap of $\frac{3}{16}$ in. at the other end. Examination of the joint shows that there is an interval of about $\frac{3}{4}$ in. before electrical contact is made between the electrode and the nipple threads. The effect is to provide a preferential path for the current through the electrode collar for at least $\frac{3}{4}$ in. Since the cross-sectional area of the nipple is approximately equal to that of the collar, the heat generated is doubled, and this is generated in a smaller volume of material to some extent thermally insulated from the central body of the electrode. If the resistivity of the material of the collar is higher than normal, the additional heating effect will be even greater and may raise the temperature sufficiently to promote the more rapid oxidation sometimes met in practice.

A better distribution of current between collar and nipple might be obtained by the use of tapered nipples, a regular practice in the U.S.A. The gap between the end of the nipple and the electrode then occurs where the cross-section of the nipple is small and that of the collar large. Additional advantages of this type of joint are that the time for screwing on a new electrode length is reduced and the mechanical strength is increased. Excessive wear is not likely to be encountered when the resistivity of the collar material is normal; in every case where trouble has occurred and the resistivity has been measured, the values have been high.

Table IV

ELECTRICAL RESISTANCE OF 1-ft. LENGTHS ALONG TWO ELECTRODES JOINED WITH A NORMAL NIPPLE

Successive 1-ft. Lengths Along Combined Electrodes	Current, amp.	Potential Difference, mV.	Resistance of 1-ft. Length, microhms
1	10	0·61	61
2	,,	0·60	60
3	,,	0·62	62
4	,,	0·66	66
5 (containing joint)	,,	1·32	132
6	,,	0·64	64
7	,,	0·57	57
8	,,	0·55	55
9	,,	0·54	54

NON-DESTRUCTIVE TESTS ON COMPLETE ELECTRODES

Following the preliminary investigations, non-destructive resistivity tests were applied to complete electrodes. Two types of tests were used: one is the standard method for measuring low resistances, *i.e.*,

Table III

ELECTRICAL RESISTIVITY TESTS ON COMPLETE 10-in. dia. ELECTRODES

Electrode No.	Length, in.	Weight, lb.	Current, amp.	Potential Difference Across 3 ft., mV.	Resistance of 3-ft. Test Length, microhms	Equivalent Electrical Resistivity, microhms/c.c.
1	57	291	10·0	1·72	172	960
2	60½	299	9·6	1·82	190	1040
3	68	283	9·5	1·84	194	1070
4	70¼	294	9·8	1·81	185	1020
5	67¾	291	9·5	1·61	170	940

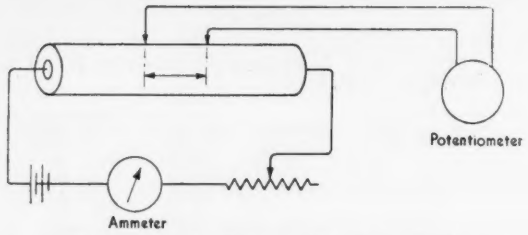


Fig. 7—Circuit diagram (direct method)

by measuring the potential fall over a measured length of the specimen when a known current passes, and the other is the core-loss comparator.*

Direct Method

The simple circuit diagram is shown in Fig. 7. A D.C. current is passed through the electrode and the potential fall over a measured length is measured by means of two probes connected to a potentiometer. The resistivity in ohms/c.c. is Vr^2/Il, where V = voltage drop; I = current, amp.; l = measured length, cm.; and r = radius, cm.

The arrangement used by Heselwood is shown in Fig. 6. The current connections are made through ordinary electrode nipples screwed into the ends of the electrode under test, each of the nipples having a small brass terminal screw set into one of its ends, and the potential tappings are taken by means of pointed brass rods mounted on insulating rod. Some typical results using this method are given in Table III, and Table IV shows the electrical resistance of successive 1-ft. lengths of two electrodes jointed with a nipple. The resistance of the section including the joint is double that of adjacent sections on either side.

The arrangement used by Thomas is shown in Fig. 8. The probe was provided with four points: the outer ones supplied the D.C. current and the inner pair measured the potential drop. Laboratory checks on this method showed that it gave somewhat high figures; this may be connected with the fact that with the arrangement described the current distribution in the portion under test is not likely to be uniform. The method is, however, considered to be sufficiently accurate for comparison purposes.

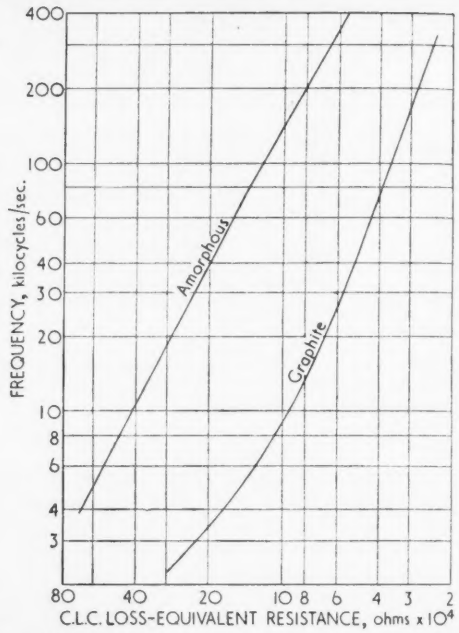


Fig. 9—Core-loss comparator loss expressed as equivalent resistance over 2·8–160 kilocycles/sec. frequency range for an amorphous and a graphite electrode

Core-Loss Comparator

This apparatus,† originally developed in the Bragg Laboratory in Sheffield, consists essentially of an unstable oscillator, the output of which after amplification may be viewed on a cathode-ray tube.

The specimen is placed within, or in contact with, a coil which is so connected that it will affect the

* The use of this apparatus for measuring resistivities was suggested by Dr. J. R. Rait.

† The adaptation of the core-loss comparator for measuring resistance of electrodes was reported to the Sub-Committee by Dr. J. R. Rait and by Dr. M. D. Jepson and Mr. R. Lomax of Hadfields Ltd.

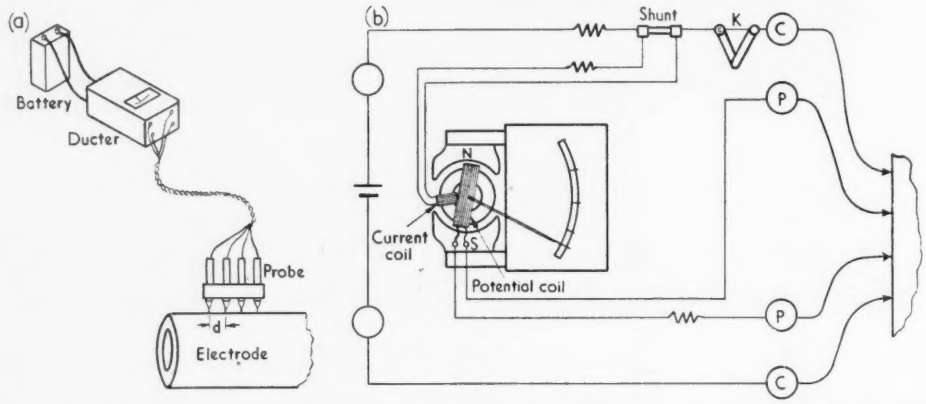


Fig. 8—Apparatus for measuring specific resistance of graphite electrodes, using the Evershed 'ducter' ohmmeter

Table V

EXAMINATION OF ¾-in. dia. SPECIMENS

(Hadfields Ltd. and United Steel Companies, Ltd.)

Specimen	C.L.C.—Equivalent Resistance, ohms	Resistivity microhms/c.c.
Good electrode	40,000	Not measured
A	47,000	1020
B	56,000	1320
C	75,000	1450
D	82,000	1750

amplitude of the oscillator without changing its frequency appreciably. The change in amplitude is dependent on the hysteresis and resistivity losses in the coil caused by the specimen. To obtain a numerical basis for comparison the cathode-ray tube is replaced by a valve voltmeter. For each specimen the oscillator is adjusted to give an arbitrary reading on the voltmeter; a decade resistance box is then connected across the coil and adjusted to obtain the chosen reading on the voltmeter. The induction losses can then be expressed as equivalent ohms on the resistance box. Since the instrument can work at any frequency from 1 to 250 kilocycles/sec., materials which cannot be sorted at one frequency may be sorted at another. The depth of penetration varies with the frequency and with the material under examination; in the case of graphite the range is about 10–0·5 cm. over the frequency range 2·8–160 kilocycles/sec.

As an initial test a sample from a good electrode was compared with a sample of amorphous carbon; the results (*see* Fig. 9) were widely different over the full frequency range used, *i.e.*, 2·8–160 kilocycles/sec. The wide difference in induction losses made it appear probable that sorting different degrees of graphitization could be accomplished, and this was confirmed

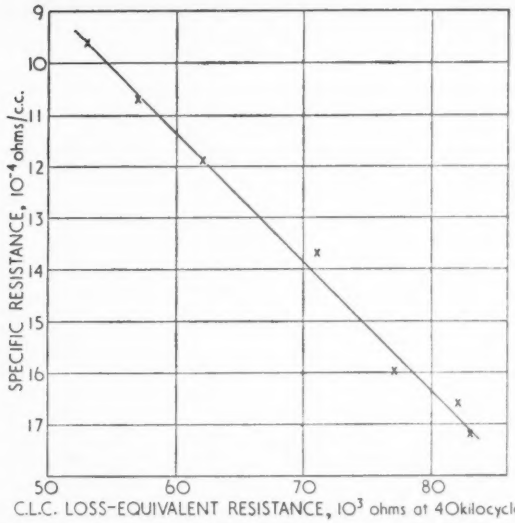


Fig. 10—Core-loss comparator loss *vs.* specific resistance for middle 1-ft. length of electrode

Table VI

RESULTS ON 9-in. dia. ELECTRODES

(Hadfields Ltd. and United Steel Companies, Ltd.)

Specimen No.	C.L.C.*	Resistivity, microhms/c.c.	Specimen No.	C.L.C.*	Resistivity, microhms/c.c.	Specimen No.	C.L.C.*	Resistivity, microhms/c.c.
1	53	960	34	54		67	57	
2	57		35	64		68	57	
3	71		36	71		69	57	
4	52		37	54		70	68	
5	57	1070	38	50		71	54	
6	71	1370	39	55		72	60	
7	55		40	60		73	52	
8	62	1195	41	51		74	55	
9	59		42	55		75	59	
10	59		43	62		76	52	
11	58		44	55		77	54	
12	57		45	64		78	59	
13	76		46	52		79	49	
14	82	1680	47	54		80	48	
15	83	1730	48	79		81	51	
16	64		49	54		82	54	
17	57		50	51		83	58	
18	54		51	57		84	58	
19	60		52	51		85	68	
20	57		53	55				
21	59		54	63		Screws		
22	62		55	48				
23	77		56	56		1	56	
24	86		57	57		2	51	
25	59		58	52		3	54	
26	60		59	56		4	51	
27	61		60	56		5	56	
28	75		61	64		6	49	
29	57		62	64		7	57	
30	65		63	52		8	63	
31	64		64	60				
32	75		65	62				
33	49		66	69				

* Core-loss comparator, equivalent resistance in ohms × 10^3, using 40-kilocycles/sec. coil

Table VII

RESULTS ON 12-in. dia. ELECTRODES

(Hadfields Ltd. and United Steel Companies, Ltd.)

Specimen No.	C.L.C.	Resistivity, microhms/c.c.	Specimen No.	C.L.C.	Resistivity, microhms/c.c.	Specimen No.	C.L.C.	Resistivity, microhms/c.c.
1	58		18	66		35	62	
2	57		19	67		36	53	
3	63		20	69		37	57	
4	77	1400	21	66		38	51	
5	59		22	59		Screws		
6	60		23	64		1	55	
7	77	1470	24	57		2	52	
8	65		25	51		3	51	
9	64		26	68		4	49	
10	67		27	53		5	49	
11	58		28	59	1220	6	53	
12	75	1470	29	59	1200	7	52	
13	59		30	58	1120	8	49	
14	60		31	67	1270	9	54	
15	72	1420	32	64	1070	10	51	
16	65		33	59	1170	11	56	
17	70	1500	34	59		12	51	

when rod specimens, covering a range of resistivities, were examined.

It was known that working with equivalent resistances much greater than 100,000 ohms led to unstable conditions and made measurement difficult. The curves in Fig. 9 show that a coil working at a frequency greater than 10 kilocycles/sec. and preferably to the order of 100 kilocycles/sec. was required. However, it is advisable to avoid very high frequencies since, as already mentioned, the penetration decreases rapidly with increasing frequency and the instrument becomes more sensitive to surface defects. As a compromise it was decided to use coils working at 40 kilocycles/sec., having the largest overall diameter compatible with strength, further to minimize interference from small local defects.

Four rod specimens, one longitudinal and one transverse from each of two electrodes, were tested by inserting them in the coil and measuring the equivalent resistivities; the results, with the specific-resistance values determined by the United Steel Companies, are given in Table V. The Table shows that there was good agreement between the core-loss comparator values and those determined by the standard method, but they were obtained on $\frac{3}{4}$-in. dia. rod and not on full electrodes.

Tests were also made on complete 12-in. electrodes. The test coil was held against the end of the electrode at the base of the thread while the oscillator was adjusted and was then held away from the electrode while the equivalent resistance was obtained. In all, 123 electrodes and 20 screws were tested, each test taking on an average about 2 min. The results are given in Tables VI and VII, which also include for comparison some resistivity values measured over the mid-length of a few selected electrodes. In Fig. 10 the induction losses are plotted against the specific resistance over the middle 1-ft. length of the electrode; a reasonably linear relation holds, at least between 10×10^{-4} and 17×10^{-4} ohms/c.c. This graph relates the induction losses, as measured by the core-loss comparator, at the end of the electrode to the average specific resistance over the middle 1 ft. Unless the resistivity of the electrode is uniform along its length this cannot be a direct relationship; it is, however, a convenient way of obtaining an indication of the average resistivity of the electrode.

The results obtained on the 12-in. electrodes were in general higher than those obtained on the 9-in. electrodes, although the spread was not quite so wide. Electrode 38 in Table VII was from an old stock (1948) and was known to be good, this being confirmed by the very low comparator reading.

It was decided to test the variation of resistivity along the two worst electrodes, *i.e.*, 14 and 15 in Table VI. Because the coil had to be held on the curved surface of the electrodes, the results were found to be less reproducible. Comparator readings were taken at five positions, 1 ft. apart, along the

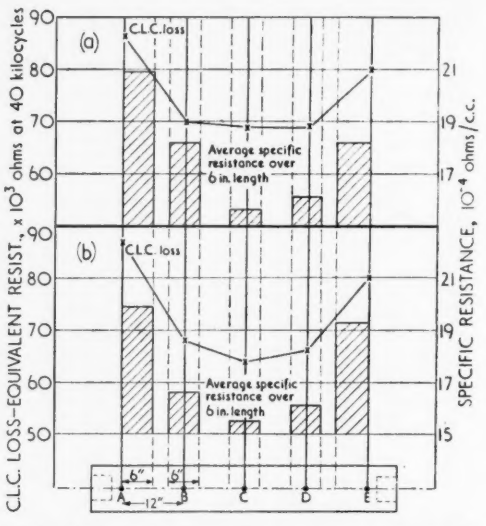


Fig. 11—Variation of core-loss comparator loss and specific resistance along electrode

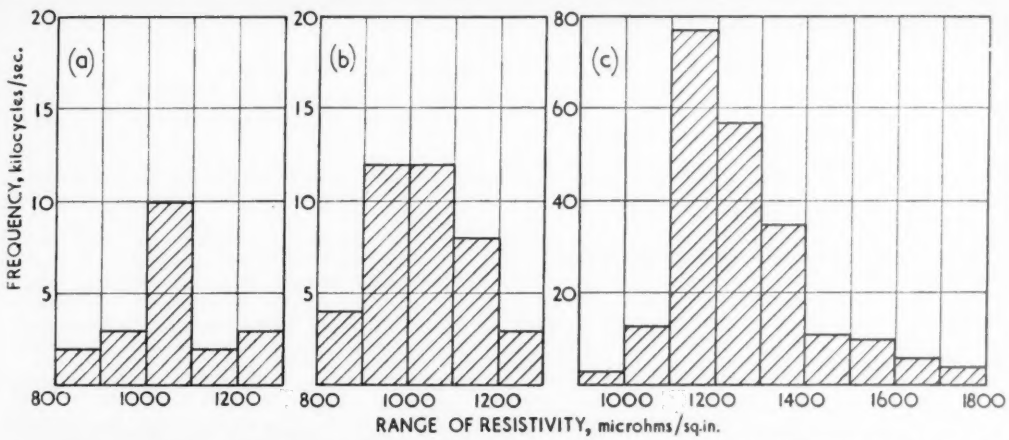


Fig. 12—Histogram showing variation in resistivity in electrodes used by: (*a*) Clyde Alloy Steel Co., Ltd. (20 8-in. dia. electrodes); (*b*) B.I.S.R.A. (39 4-in. dia. electrodes); (*c*) Stewarts and Lloyds Ltd. (216 16-in. dia. electrodes—ends only)

length of the electrodes, and the specific resistances were measured on 6-in. gauge lengths. At each end of the electrode, owing to the screwed recess, the resistance had to be measured over the 6 in. inward from the test points, but at the other positions it was measured between points 3 in. on either side of the comparator test point. This is shown diagrammatically with the results in Fig. 11. The resistance of the electrodes is much higher at the ends than in the middle, and this explains some of the anomalous results when comparing the comparator results with the resistivity values. Similar tests on other electrodes showed that in general the high-resistance electrodes were much worse at the ends than in the middle, but the better electrodes showed very little variation. Because the core-loss comparator can measure resistivities in localized areas it is more suitable than the other method for routine acceptance testing.

ROUTINE NON-DESTRUCTIVE TESTING

As a result of this preliminary work, a number of firms now test all incoming electrodes as a matter of routine; others use the test only when trouble is suspected. Owing to delays in obtaining the necessary equipment to construct a core-loss comparator, the direct method, illustrated in Figs. 6 and 7, has been most widely used. Some typical figures are shown in Table VIII, and histograms giving the variation in resistivity of electrodes used in three plants are given in Fig. 12. This Figure shows that the highest proportion of the samples falls within the recommended range of resistivities, but that there are always some which have unduly high values. There is also a tendency for resistivity to be higher at electrode ends, shown both by the peak distribution occurring in a slightly higher range and by the distributions being markedly skew.

Table VIII

GRAPHITE-ELECTRODE RESISTIVITY

J. Nicol (Clyde Alloy Steel Co., Ltd.)

Rod No.	Resistivity, microhms/c.c. End	Middle	End
1	1310	1240	1440
2	940	860	910
3	1030	1030	1100
4	1470	1210	1210
5	1090	1060	1060
6		1020	
7		1070	
8	1250	1170	1250
9	1150	1150	1200
10		1000	
11		1030	
12		1060	
13		880	
14		980	
15		1070	
16	1220	1220	1220
17		1010	
18		1000	
19		1060	
20		1030	

CONCLUSIONS

(1) Spectrographic examination of a sample from a 'bad' electrode showed it to be of high standard of purity. It therefore seems that poor quality is not due to the use of unsatisfactory materials.

(2) Abnormal wear at the collar of an electrode is usually associated with a high electrical resistivity. X-ray examination suggests that this is due to incomplete graphitization. A method has been developed and adopted for the non-destructive testing of complete electrodes.

(3) Resistivity tests on complete electrodes show that the ends of a 'bad' electrode have a higher resistivity than the middle. For this reason the core-loss comparator has an advantage over the standard method of measuring low resistances in that it can measure resistances in localized areas.

References

1. C. Hering: *Transactions of the American Electrochemical Society*, 1909, vol. 16, pp. 265–316.
2. F. N. Pensa: Proceedings of the 3rd International Congress on Electroheat, 1953 (in the press).
3. H. Davis and H. C. Hottel: *Ind. Eng. Chem.*, 1934, vol. 26, pp. 889–892.
4. C. M. Tu, H. Davis, and H. C. Hottel: *Ibid.*, 1934, vol. 26, pp. 749–757.

Correspondence on the Paper—

THE FLUIDITY OF MOLTEN STEEL*

By B. G. Rightmire and H. F. Taylor

Mr. **R. W. Ruddle** (British Non-Ferrous Metals Research Association) wrote: Ruff's failure to analyse correctly his experimental results† has led to some unfortunate confusion. Rabinovich‡ was the first to point out the errors in Ruff's treatment of his results but, as Professor Rightmire and Professor Taylor state, the equation used by Rabinovich in his re-analysis of Ruff's data is itself incorrect, and they have therefore provided a new analysis of Ruff's results. Unfortunately, this new treatment is itself not entirely satisfactory, for the following reason.

The authors, in their Appendix I, quite correctly apply the usual steady-flow energy equation to Ruff's experiments, obtaining the pair of equations:

$$x_1^2\left(1 + K + \frac{fl_1}{d}\right) = 4hH \quad \text{.....(i)}$$

$$x_2^2\left(1 + K + \frac{fl_2}{d}\right) = 4hH \quad \text{.....(ii)}$$

where the notation is that given in Fig. 2 and in the text of the paper. They then eliminate hH between these equations, obtaining the expression:

$$f = \frac{(x_2^2 - x_1^2)\,(1 + K)d}{x_1^2 l_1 - x_2^2 l_2} \quad \text{.....(iii)}$$

Next they use equation (iii) to calculate values of the friction factor f from Ruff's experimental results, assuming that $K = 0$. They state (p. 168) that K is a coefficient to account for losses due to turbulence at the tube entrance. They add that for a sharp entrance K may be as much as 0·5, and since, in equation (iii), f is proportional to $(1 + K)$, it is quite clear that the values of f given in their Table I may be too low by as much as $33\frac{1}{3}$%.

The authors, in their treatment, seem to suggest that the only loss of head in Ruff's apparatus, apart from that due to wall friction in the horizontal tubes, is the entrance loss mentioned above. In fact, there must almost certainly have been a loss in the vertical part of the system, as Ruff's original drawing shows. This point is perhaps of small importance, for such losses may be regarded as being incorporated in K. They are mentioned, however, to make the following analysis quite clear.

The first part of my own treatment is identical with that of Rightmire and Taylor, and leads to the derivation of equations (i) and (ii). Instead, however, of eliminating hH from these equations, I prefer to eliminate the unknown K, since both h and H are given in Ruff's original paper. The following equation is then obtained:

$$f = \frac{4hHd}{(l_2 - l_1)}\left(\frac{1}{x_2^2} - \frac{1}{x_1^2}\right) \quad \text{.....(iv)}$$

Applying equation (iv) to the data assembled in the authors' Table I, the values for the friction factor f given in Table *A* are obtained. These values are, in general, greater than those calculated by the authors, in some cases by as much as 35%.

The accuracy of the friction factor thus evaluated depends, *inter alia*, upon the extent to which H, the height of the metal in the vertical part of the system, was maintained constant in Ruff's experiments. Ruff mentions that a flow-off was used to keep the height of the metal in the funnel constant; therefore, provided that Ruff took steps to prevent the velocity head of the stream of metal from the ladle from adding substantially to the head in the vertical system, the factors given in Table *A* should be reasonably accurate and certainly ought to be better than those quoted in the authors' Table I. However, the fact that in two instances the friction factors calculated from equation (iv) are lower than the corresponding figures derived by Rightmire and Taylor, using equation (iii), suggests that in Ruff's experiments the head was not adequately controlled and in some cases fell below the nominal value.

I have given the above analysis in the hope of finally clarifying the interpretation of Ruff's results, and have no wish to criticize this otherwise admirable paper.

AUTHORS' REPLY

Professor **B. G. Rightmire** and Professor **H. F. Taylor** wrote in reply: The authors are grateful to Mr. Ruddle for his discussion of the friction factors obtaining in Ruff's experiments. The fact that his more accurate computations yield higher values of f than those of the authors supports the conclusion that the flow in Ruff's experiments was turbulent.

* *J. Iron Steel Inst.*, 1953, vol. 175, pp. 167–176.

† W. Ruff, *Iron and Steel Institute, Carnegie Scholarship Memoirs*, 1936, vol. 25, pp. 1–39.

‡ Y. Z. Rabinovich, *Doklady Akademii Nauk S.S.S.R.*, 1946, vol. 54, No. 5, p. 391.

Table A

COMPARISON OF FRICTION FACTOR OBTAINED BY TWO METHODS

Metal	x_1, cm.	x_2, cm.	l_1, cm.	l_2, cm.	d, cm.	f From Equation (iv)	f From Equation (iii)
Iron	30·6	25·1	5	15	0·5	0·039	0·033
	29·0	25·4	5	15	0·8	0·043	0·028
	32·0	27·1	5	15	0·6	0·035	0·030
	22·2	18·8	5	15	0·5	0·028	0·025
	23·4	16·8	5	25	0·5	0·030	0·031
Steel	25·9	22·2	5	15	0·8	0·030	0·035
	32·2	27·5	5	15	0·8	0·043	0·037
	38·4	32·6	5	15	0·8	0·046	0·039

The British Iron and Steel Research Association

SHEFFIELD LABORATORIES

Introduction

THE policy of the Association since its formation in 1945 has been to establish laboratories in different centres of the steel industry to serve both local firms and the industry as a whole. The Sheffield Laboratories, which were officially opened by H.R.H. the Duke of Edinburgh on 19th November, 1953, house most of the Association's laboratories for the making, mechanical working, and metallurgy of steel. They include two 'plant laboratories' in which the results of research can be tried out on an intermediate scale before they are applied to full-scale production plant.

Each of the three Divisions (Steelmaking, Mechanical Working, and Metallurgy) is responsible for its programme and progress to its own section of the industry through the appropriate B.I.S.R.A. Divisional Panel, which is composed mainly of representatives from the industry. Each Division has its own laboratory and can make use of the special equipment required for its research programme, but general research facilities are provided through a central engineering section, to avoid duplication of equipment. A local group manager co-ordinates the needs of the three laboratories and is responsible for all the central administration.

The new laboratories, shown diagrammatically in Fig. 1, comprise a main laboratory and office block, a metalworking shop, and a melting shop. In the main block, all the internal walls are easily removable to allow new layouts to be built. The main services (electricity, gas, water and drainage, heating, and air-conditioning) are carried in four vertical brick-built shafts from the basement to a comprehensive and easily accessible system of horizontal ducts running in false ceilings to the corridors and rooms. With this system, offices can be converted into laboratories and *vice versa* without any structural alterations; the loss of floor space due to the ducting is less than 1% of the total, and the wall space is entirely unaffected. The metallurgical laboratories contain facilities for heat-treatment, metallurgical examination, and mechanical, magnetic, and electric testing, as well as for X-ray structural studies and chemical, spectrographic, and vacuum-fusion analysis.

The single-storey metalworking shop adjoins the main block, but its foundations and structure are completely separate, to minimize the transmission of vibration from the heavy plant. The latter comprises a strip rolling mill, a forging press, and a wire-drawing machine, together with some smaller equipment and a 10-ton electric overhead travelling crane.

The melting shop houses a 10-cwt. electric arc furnace and the furnace transformer.

THE METALLURGICAL LABORATORIES

The Metallurgy Division was set up to extend the work previously carried out by the three Joint Committees (Alloy Steels Research, Corrosion, and Standard Methods of Analysis) of The Iron and

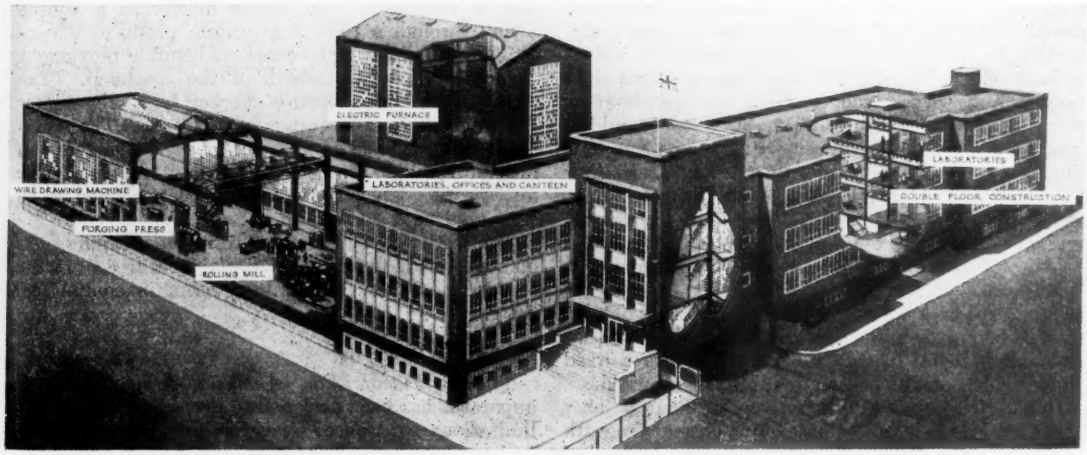


Fig. 1—Layout of Sheffield Laboratories

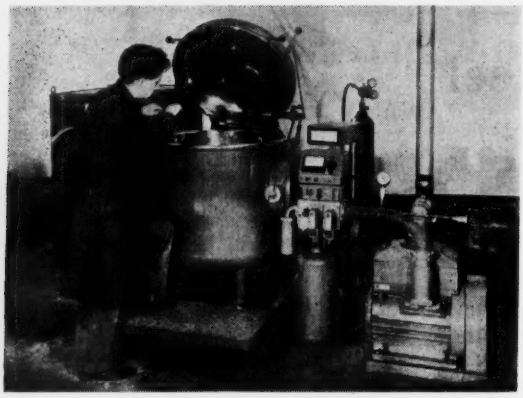


Fig. 2—Vacuum-melting induction furnace

Steel Institute and the British Iron and Steel Federation, reporting to the Iron and Steel Industrial Research Council. Two other Committees (Metal Physics and New Techniques) were added, to carry out research on fundamental problems affecting the technology of iron and steel. The Metallurgical Laboratories at Sheffield are concerned with the effects of composition, structure, and treatment on the physical and mechanical properties of steels.

Research on Pure Materials

A number of the Division's researches are being carried out on pure materials. In recent years the National Physical Laboratory (Metallurgical Division), in collaboration with B.I.S.R.A., has developed a method for the manufacture of iron of high purity, *i.e.*, having total impurities of not more than 0·027%. The H.F. vacuum furnace and other ancillary equipment in the Division's laboratories will enable iron of similar purity to be made there. This equipment (*see* Fig. 2) comprises a high-vacuum melting and casting set (55 lb. capacity) and an open-air melting furnace (100 lb. capacity), both of which are induction-heated. The former consists of a fixed double-walled stainless-steel container with a rotatable coaxial H.F. lead-through forming a carrier for the induction coil holding the melting crucible. The container is water-cooled, and the hinged lid is fitted with a quartz glass window and with a vacuum-tight storage chamber which holds up to 40 lb. of material to be added during the melt. A quick-immersion thermocouple is operated by a handle outside the container. Melting can be carried out in an inert gas atmosphere under reduced pressure if necessary. The high-vacuum connection has a large cross-section, to enable low final pressures to be obtained. In cold metal practice a pressure of 3×10^{-5} mm. of Hg can be obtained in 30 min., and with hot metal 10^{-4} mm. of Hg is obtainable. When casting, the induction coil is tilted so that the molten metal in the crucible pours into a fixed mould which can be heated or cooled as required.

In addition to providing pure iron for the researches described below, it will be possible to extend the work which has been carried out at the National Physical Laboratory (with the encouragement of the Alloy Steels Committee) on the effect of alloying elements on the properties of iron and steel. Such background information will assist the industry in the development of steels for special purposes.

Strain Ageing—A fundamental study is being undertaken of the strain-ageing characteristics of pure iron, with the object of obtaining background information on the mechanism by which carbon and nitrogen affect them. Unlike previous work on the subject, this involves study of the independent effect of carbon and nitrogen in samples of high-purity iron initially free from these elements. Further work will include study of the influence of other elements, including V, Al, and Ti, with the object of assisting the development of non-ageing rimming steels.

Silicon Iron—Experiments are in progress to discover the effect of impurities on the *B–H* characteristics of pure silicon-iron cold-rolled sheet. From measurement of the magnetic properties by a specially designed torque magnetometer, an approximate indication of the crystallographic texture of samples of silicon iron is obtained. More accurate determinations of texture in selected samples are obtained by orthodox X-ray methods. Accurate determinations of the *B–H* characteristics of the iron are made on a commercial permeameter.

Tool Steels—The problem of the manufacture of large ingots of high-speed steel free from coarse segregates of carbide has still not been solved, and methods of controlling the freezing of steel ingots are to be investigated.

Fatigue Properties of Steels

Previously no systematic investigation has been undertaken to determine the magnitude of the effect of non-metallic inclusions on the fatigue properties of engineering steels. A programme of research has therefore been started on a commercial cast of three typical steels. The fatigue properties of material from the bottom, middle, and top of typical commercial casts of these steels are being studied. Twelve fatigue-testing machines of the 2-point loading, rotating bending type have been installed in the fatigue laboratory. In these machines, a couple acting through a rigid link system produces a condition approximating to constant bending movement over the whole test length of the specimen. The diameter of the specimen is checked by a micrometer arrangement to an accuracy of 0·0001 in. The maximum speed is 5000 r.p.m., so that each machine can provide 7 million stress cycles per day. A comparator microscope enables the finish of each specimen to be compared with a standard.

The work will be extended to include fatigue tests on steels made from pure materials and free from the segregations that are usually present in commercial casts.

Steelworks Analysis

Vacuum-Fusion Gas Analysis—A gas analysis apparatus has been designed to overcome some of the limitations experienced by earlier users of the vacuum-fusion method. The apparatus consists of an induction-heated graphite crucible into which a solid sample is

introduced, through a greaseless vacuum lock, in an evacuated system at 10^{-6} mm. of Hg and at 1700° C. A high-speed mercury vapour diffusion pump capable of operating against high backing pressure is used to remove the gaseous products of reaction. These are collected in a Toepler pump and transferred to gas analysis apparatus.

Analysis of Steelmaking Materials—In view of the increasing variations in home and foreign iron ores, and the presence in them of major contents of Ni, Cr, and Al compounds, the routine analytical procedures are being checked. Decomposition procedures for blast-furnace raw materials, slags, and steelmaking additions have been studied with a view to obtaining better control of steel quality, and rapid and reproducible methods for the determination of sulphur have been developed. Solution methods of spectrographic analysis are being investigated, and a traversing apparatus is being built for the analytical profiling of weld areas and segregates and the study of diffusion mechanism. Future research will include chromatography, polarography, and micro-techniques, and investigations of the fundamental principles of mercury-cathode electrolysis and of the mechanism of reduction of a sulphate solution, in an attempt to determine sulphur as hydrogen sulphide.

THE STEELMAKING LABORATORY

The Steelmaking Division was set up to extend the work previously carried out by the two Joint Committees (Open-Hearth and Heterogeneity of Steel Ingots) of The Iron and Steel Institute and the British Iron and Steel Federation, reporting to the Iron and Steel Industrial Research Council. The Division occupied temporary quarters in the Swansea and Battersea Laboratories until 1949, when part of the Sheffield buildings became available. It is concerned mainly with the design and operation of open-hearth and electric furnaces, the process chemistry of steelmaking, and ingot casting. The researches on ingots and the electric arc furnace are being carried out by the same section, as the arc furnace supplies much of the steel for the ingots.

Electric Arc Furnace (*Fig.* 3)—This 10-cwt. furnace is of modified Héroult design, incorporating several special features. The shell is 5 ft. dia. and the height from sill to roof ring can be varied by inserting separate sections. The maximum and minimum heights are 2 ft. 2 in. and 1 ft. 5 in., respectively. The electrodes in use at present are standard 4-in. dia. graphite on a 12-in. dia. pitch circle, and the latter is variable up to the full diameter of the bath. The furnace bottom is of austenitic steel, to permit induction stirring equipment to be fitted if desired. The maximum forward tilt of the furnace is 60° and the reverse tilt is 15°. The furnace transformer has an exceptionally large nominal capacity (500 kVA.), so that the effect of high power-input rates can be investigated.

Process Chemistry

The principle aim of this section of the Steelmaking Division is the application of results of fundamental work to the solution of problems arising in production.

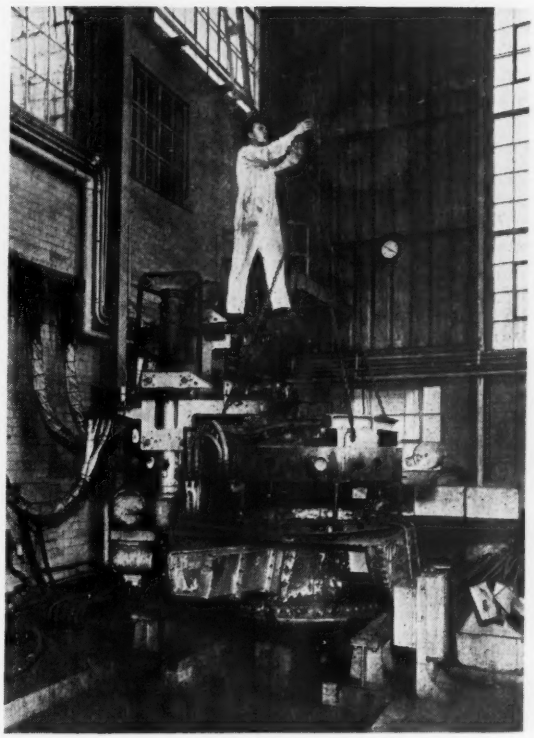


Fig. 3—10-cwt. electric arc furnace

Sulphur Removal—Extensive investigations are being carried out in connection with desulphurization in open-hearth furnaces and pre-open-hearth refining treatments. It has been found that the rate of desulphurization depends mainly on the extent of agitation in the mixer, and various methods of slag/metal mixing have accordingly been studied. These include atomizing the molten iron, pouring it directly into the slag, pouring it through a carbon 'colander' so that it forms fine jets into the slag, and blowing the slag/metal mixture with nitrogen, through a carbon probe, at various pressures. A detailed study of the soda-ash process has shown that it is still the most practicable method of desulphurizing large tonnages of iron.

Model Chemical Systems—Two small-scale 'cold' models of slag/metal systems have been constructed, in an attempt to overcome the practical difficulties involved in the above 'hot' mixing experiments. Preliminary experiments were carried out on a system in which the metal was represented by sodium amalgam and the slag by aqueous sulphuric acid, and in which the mixing efficiency was measured by the rate of sodium transfer to the aqueous layer. Mixing techniques investigated include blowing gas through a single probe at various depths, and through several holes in the bottom of a container; using a stirring rabble; and pouring the 'metal' into the 'slag' from various heights and at different rates.

In the second model system the metal is represented by an aqueous solution of zinc chloride, potassium

iodide, and iodine, and the slag by *n*-hexane; the rate of iodine transfer gives a measure of the mixing efficiency. The materials used in this system are relatively inexpensive, and it can therefore be used on a much larger scale than the first system.

Open-Hearth Furnace Operation

The model furnace at the Sheffield Laboratories has been designed to operate at steelmaking temperatures. Works' investigations have been made of the influence of reversal time and of air infiltration on preheat, and attempts are being made to develop longer-lasting shields for suction pyrometers. A periscope which scans and photographs the furnace roof from inside the furnace, under operating conditions, is being tested, and a ' tracer ' technique, using radon, has been developed as an alternative to the use of Pitot tubes for measuring gas flows. Works trials are to be carried out of a new oil burner, developed by the Physics Department, to determine the maximum momentum efficiency.

The B.I.S.R.A. Steelmaking Laboratory is to make field trials of the effect of furnace operating conditions on flame heating power, to extend the experimental work carried out at Ymuiden by the International Flame Radiation Committee.

Electric Steelmaking

Research is in progress on the power consumption and optimum power input of the experimental 10-cwt. furnace (Fig. 3). Power inputs greater than 500 kW. have been applied with increasing efficiency up to the maximum input used. The refractory wear and overall heat balance in the furnace are also being investigated.

A series of experiments is being carried out to investigate electrode wear and variations in quality, the use of protective electrode coatings, and the optimum electrode diameter. Future work will include trials of Söderberg electrodes.

Ingot Casting

A study has been made of the factors affecting the structure of balanced steels, particularly the type, size, and distribution of sub-surface blowholes. It was found that the distribution is distinctly related to the state of oxidation of the steel as it enters the mould, but not to that as it leaves the furnace. Accordingly, works trials are being carried out to determine the losses of the more easily oxidizable elements from the steel between furnace and mould, and whether these are due to air oxidation or to oxidation by the slag.

Solidification—The effect of the rate and manner of solidification on the structure of large and small ingots is being studied. Further work is to be done on the design of octagon moulds to prevent ingot corner segregation, and on the occurrence of sub-surface pinholes in killed steel.

Moulds—The normal mechanical properties of mould metal do not seem to be related to the performance of the mould, and tests are to be made of other properties, such as damping capacity and resistance to growth. The function of mould dressings is also being investigated. A method has been developed by which the steel rising in a small mould can be viewed through a clear silica window in the side of the mould and photographed at high speeds. The performance of large moulds and foundry techniques for mould making are also being studied.

Feeder Heads—Trials are being carried out on insulating refractory materials for lining feeder heads. Measurements of the heat transfer from the steel in the head to the brickwork show the efficiency of insulation. The use of refractories on large ingots is being studied in the laboratory.

Continuous Casting—A pilot plant for the continuous casting of ingots has now been designed and is shortly to be constructed at the works of Wm. Jessop and Sons Ltd. Preliminary work on the holding of steel in the ladle and the provision of a suitably heated tundish is in progress in the Sheffield Laboratory, using steel supplied by the 10-cwt. arc furnace.

THE METAL WORKING LABORATORY

The Mechanical Working Divisional Panel was set up to extend the rolling mill researches formerly carried out by the Rolling Mill Research Sub-Committee of the Iron and Steel Industrial Research Council to other branches of industrial metalworking. The Metal Working Laboratory at Sheffield studies the processes from the ingot to the finished product of plate, sheet, sections, forgings, and wire. There is also a team engaged in research on theoretical plasticity.

Principal Equipment

Rolling Mill (*Fig.* 4)—This is a high-speed, reversible, 4-high 14-in. cold strip mill. The total design load is 300 tons. The ingoing strip speed is variable from 0 to 1750 ft./min., and any speed within these limits can be preset between 10 and 30 sec. A wide range of back and front tensions can be applied, and selected values can be maintained constant during acceleration, deceleration, and change in coil diameter. The maximum tension available under normal running conditions is 2 tons per side, although this can be

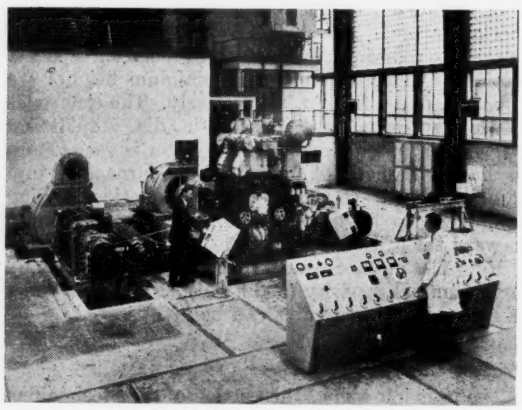


Fig. 4—4-high 14-in. cold rolling mill and control console

increased by up to 100%. Under stalled conditions the tension is automatically reduced, but provision is made for obtaining the running tension for short periods. Also, strip can be reeled from one coiler to the other under full tension control with the rolls lifted. Instruments for measuring the roll force, tensions, and strip gauge are displayed on the mill housing. The mill is equipped for research on several systems of automatic gauge control.

'Orowan' Mill—This small 2-high mill was transferred from the Cavendish Laboratory, Cambridge University, to the Sheffield Laboratories in 1950, and has only recently been installed in the metalworking shop. It is equipped for research on the hydraulic roll-loading method of automatic gauge control.

Wire-Drawing Machine (*Fig.* 5)—This is a 2-hole straight-line machine with individually motor-driven feed and take-up spools and 18-in. dia. capstans. The capstans and dies are water-cooled from a constant-pressure supply. The maximum inlet speed is 2500 ft./sec. and the maximum finishing speed is 5000 ft./sec. The full feed spool can be stopped in 2 sec. by means of dynamic braking and twin electromagnetic brakes. The machine is used to draw wire of initial diameter 0·040–0·104 in., at temperatures from 10° to 130° C. and with a reduction in area of up to 45% per pass. The entire drawing system can be immersed in lubricant, or the capstans, dies, and wire can be sprayed, the temperature of the lubricant being thermostatically controlled. Alternatively, soap boxes can be fitted for dry drawing. The interpass and spool tensions are variable between zero and the breaking point of the wire. If the wire breaks, or if the motor overspeeds or passes maximum voltage or current, the whole plant is automatically shut down.

Forging Press (*Fig.* 6)—This is a vertical down-stroke 200-ton hydraulic machine with a double-acting piston-type ram. The maximum load can be set between 4 and 200 tons and repeated with an accuracy of $\pm 2\frac{1}{2}\%$. From 0 to 100 tons load the ram speed is variable between 0·03 and 3·09 in./sec. At heavier loads the speed is reduced to 0·03–1·54 in./sec. The ram return speed is 2·85 in./sec. The pumping equipment comprises four high-speed units, each of two plunger-type pumps and a 25-h.p. driving motor, and a smaller high-speed axial plunger pump and 4-h.p. driving motor. The required pressing speed is obtained by starting a given number of the main pumps and then exhausting part of the output of one pump to obtain a fine adjustment. When the ram is stopped, pressure is maintained in the cylinder by means of the pilot pump.

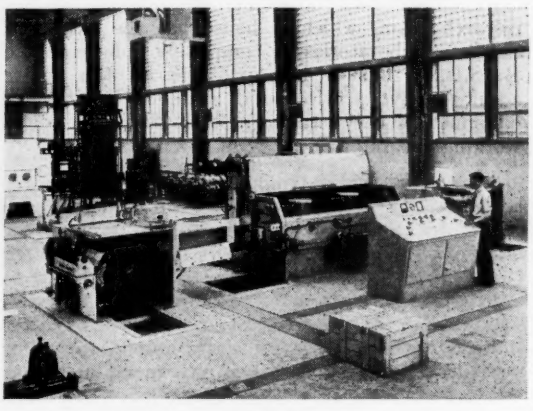


Fig. 5—Wire-drawing machine

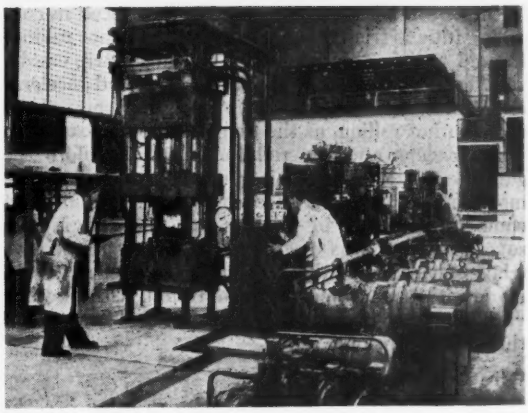


Fig. 6—200-ton forging press

A gas-fired forge furnace is used. This has a furnace chamber 4 ft. long by $2\frac{1}{2}$ ft. wide by 12–15 in. high, with viewing holes in the furnace wall. The temperature is continuously recorded by four Pt/Pt–Rh thermocouples.

Cam Plastometer—This high-speed cam-operated compression testing machine uses a 1-in. high cylindrical specimen which is compressed between two platens, one attached to the load-measuring box and the other to the moving compression table. The latter is operated by an eccentric cam and applies compressive strain at a constant rate until the height of the specimen is reduced by 50%. The speed of the cam is variable between 27 and 900 r.p.m. The maximum natural rate of strain obtainable with this cam is 40/sec., but strain rates up to 100/sec. can be obtained using different cams. The load during test is measured by a system of interference fringes projected on to recording paper mounted on a drum which is rigidly attached to the cam.

The machine is being used to obtain stress/strain curves for materials under hammer or press forging conditions. For tests at forging temperatures the specimen and anvils are first heated in a double-walled container in a furnace.

Automatic Gauge Control

Former methods of gauge control had the disadvantage that after an error in gauge was noticed there was a considerable delay before a correction could be applied. Modern roll-force meters, which are based on the use of electrical resistance strain gauges, resin-bonded on to the surface of a load-sensitive element, now enable the correction to be made immediately.

The two main causes of variation are speed in cold rolling (due to friction) and temperature in hot rolling. In cold rolling, the amount of off-gauge strip can be kept down by decreasing the acceleration and deceleration times, but this adds to the capital cost of the mill. In hot rolling, as the temperature decreases the strip hardens and the roll force and therefore the gauge increase. Automatic gauge control would give more uniform strip for cold rolling, and would overcome the problem of welding up coils differing by several thousandths of an inch in gauge. Extensive trials of various methods of gauge control are being made with the 14-in. experimental mill. Other rolling variables to be investigated are bearing clearance and thermal expansion.

Wire Drawing

The two main problems in wire drawing are the deterioration in quality of the wire with increase of drawing speed, and excessive wear of the dies. Both these effects are due mainly to the heat evolved in the die as a result of friction and plastic deformation, and methods for reducing the high temperature of the wire and for the removal of heat are being investigated.

The most important requirement is good lubrication, and improved means of applying the lubricant are being studied (*e.g.*, by electrodeposition from the solution on to the moving wire), so as to obtain more consistent lubricant film thickness and consequently more uniform die performance. Recent experiments have indicated that, contrary to general theory, relatively thick films of lubricant can be maintained in the die. (Using solid wax, a film thickness greater than 0·001 in. was obtained.) The film thickness is measured by the electric resistance of the die/wire interface. This resistance, which is much larger than would be expected with thin films, is markedly dependent on the speed of drawing. Further tests are to be made at high speeds and with radically different die profiles.

Various methods of temperature measurement have been investigated, and apparatus has been constructed for measuring the temperature of the die/wire interface close to the exit from the die, the surface temperature of the wire between die and drawing block, and the proportion of heat extracted by the die.

The wire-drawing machine (Fig. 5) has individually driven capstans, but it can be adapted to have several capstans on one shaft, all driven by one motor. It would then be practicable to use lighter drafts coupled with backpull. Work on this and other problems of machine design arising in high-speed drawing is planned.

Forging Research

A quantitative comparison (accurate to within 2%) of the behaviour of Plasticine and hot steel under geometrically similar forging conditions has shown that Plasticine at room temperature is a satisfactory substitute for steel at forging temperatures, and that the results of Plasticine experiments may be applied to full-scale forging operations.

A study is being made of the correlation between the mechanical properties of material forged in the 200-ton press (Fig. 6) at various tool ratios and reductions, and the actual forging strains as determined from geometrically similar Plasticine specimens, to determine the optimum forging reduction and best operating sequence. In practice, the columnar, equiaxed, and segregated regions of a cast ingot may behave differently under forging conditions. Tests are being made with equiaxed core material; tensile and impact properties are measured in the longitudinal and transverse directions after controlled amounts of forging strain. Future work will include the forging of small ingots.

Roll Design

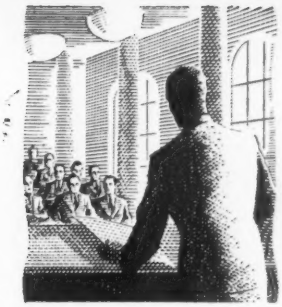
A comparative study is being made of actual pass schedules used by roll designers in Britain, with the object of determining the best method of reduction for a given section. So far, reports have been prepared dealing with the complete ranges of rounds, angles, channels, rails, and beams. Other sections to be studied are **T** bars and some special sections.

ENGINEERING DEPARTMENT

A central design, machine shop, and maintenance section is provided, under the control of a separate Group Manager, to deal with the construction of apparatus and the preparation of specimens required by the three research divisions. This Section is also concerned with the servicing, modification, and maintenance of the heavy research equipment.

CUTLERY RESEARCH LABORATORY

The Cutlery Research Council was formed in 1951 to undertake research on behalf of member firms of the Sheffield Cutlery Manufacturers' Association. The research team, which is housed at the B.I.S.R.A. Sheffield Laboratories, is completely independent of B.I.S.R.A., but it has available all the central administrative services and some of the research equipment in the metallurgy, steelmaking, and metalworking laboratories.



SPECIAL MEETING ON BORON IN STEEL

THE SPECIAL MEETING ON BORON IN STEEL, which was to have been held on 18th February, took place on Wednesday, 29th April, 1953, at the Offices of the Institute, 4 Grosvenor Gardens, S.W.1. The meeting was arranged by the Institute at the suggestion of the Ministry of Supply, acting in conjunction with the Organization for European Economic Co-operation. Mr. WILLIAM BARR, the Hon. Treasurer of the Institute, occupied the Chair.

At the MORNING SESSION, a paper by Mr. L. J. ROHL (United States Steel Corporation), entitled "The Manufacture and Use of Boron Steels in the United States," was presented and discussed, and a report of the paper and discussion follow.

At the AFTERNOON SESSION, a paper by Mr. H. B. KNOWLTON (International Harvester Company), entitled "American Applications of Boron and other Low-Alloy Steels," also reported in this issue, and a paper by Mr. R. WILCOCK (Samuel Fox and Co., Ltd.), entitled "Effect of Boron on the Mechanical Properties of Low-Alloy Steels", which appeared in the April, 1953, issue of the *Journal*, were presented and jointly discussed.

PROCEEDINGS OF THE MORNING SESSION: 9.45 A.M. to 12.30 P.M.

The **Chairman:** My first duty is to apologize for the absence of the President, who is on an important mission in the North. This Meeting originated in a proposal of the Mutual Security Agency, which came to us from the Ministry of Supply through the good offices of Mr. Oliver. I have to apologize for the absence of Mr. Oliver, who is in America; had he been here I am sure that he would be in the Chair to-day.

To the two American authors, Mr. Rohl and Mr. Knowlton, who have come so far to talk about boron and its uses in the United States, we extend a particularly warm welcome. We are also delighted that Mr. Wilcock has accepted the grave responsibility of keeping the English end up on this subject. That the importance of the subject is appreciated is evident from the very large attendance.

The Manufacture and Use of Boron Steels in the U.S.A.

By Louis J. Rohl

SYNOPSIS

After a survey of the development of boron steels as a means of conserving critical alloys, the paper describes methods of manufacture of boron steels, using simple and complex boron ferro-alloys. The effect of the type of ferro-alloy used is discussed. The paper ends with a summary of the amount of critical steel-making alloys saved by the increased use of boron. 779A

THE developments of the past few years show considerable promise in the use of boron steels, with resultant savings in the amount of critical alloys required for many applications.

Experience in the manufacture and application of boron steels in the United States has been largely in the field of constructional alloy steels, *i.e.*, those principally used for automobile, aircraft, tractor, and oil-well drilling parts requiring heat-treatment by quenching and tempering.

Before discussing the experience in the manufacture and use of boron steels in the United States, it is helpful to review the trends in the alloy content of constructional alloy steels used in the United States since 1900. Figure 1 shows the average percentages of nickel, chromium, molybdenum, and vanadium in steels specifying those elements during the period 1900–1951. Between 1900 and 1940, the average nickel content dropped progressively from 3·50 to 1·90%, and the average chromium content dropped from 1·50 to 1·05%, while the average molybdenum

Manuscript received 18th February, 1953.

Mr. Rohl is Chief Metallurgical Engineer of the United States Steel Corporation

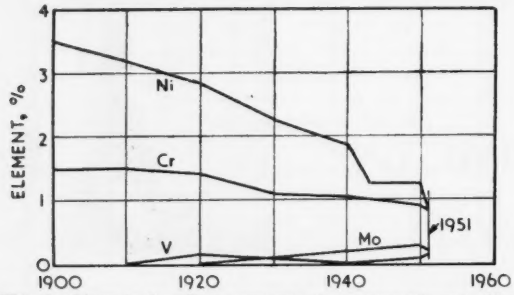


Fig. 1—Trends in alloy content of constructional alloy steels, 1900–1951

content increased from 0 to 0·18%. Some of the factors affecting these trends were:

(i) Advances in the science of metallurgy as applied to the understanding of the properties required in alloy-steel parts
(ii) The effort to economize in raw material costs
(iii) Advances in design and engineering to permit the use of steels with lower alloy contents
(iv) Improvements in heat-treating practices and a better understanding of the importance of hardenability as related to the maximum inherent properties obtainable from a combination of alloying elements.

The sharp drop in the nickel curve from 1940 to 1943 was the result of conservation efforts during World War II, which stressed the use of the National Emergency Steels containing about 0·50% of nickel, 0·50% of chromium, and 0·20% of molybdenum. The National Emergency Steels accounted for about 30% of the total production of constructional alloy steels during the war. After the war, the alloy content remained fairly constant at these lower levels until the outset of the Korean incident in 1950. The second sharp drop in the nickel curve and the downward trend in molybdenum from 1950 to 1951 were the result of restrictions in the use of these critical elements and the development of the new alternative steels, many of which contain boron.

Analysis of these curves shows that the downward trends in alloy content of U.S. steels were due to the foregoing factors, and that the effect of the war periods was to advance the times at which they occur.

About 1938 it was noticed, quite by accident, that the use of certain complex deoxidizers containing aluminium, titanium, zirconium, calcium, etc., greatly increased hardenability, and that this increase in hardenability was accompanied by a marked increase in the strength and ductility of automobile parts heat-treated by quenching and tempering. Further investigation revealed that these steels all contained boron.

During the next few years many studies and experiments were conducted to determine the effect of boron in various steels on the properties obtained. These properties were compared with those of established compositions, and the results of much of this work were published in 1942.[1]

During World War II, with the need for conservation of alloys, many thousands of tons of boron steels were made for armour, torsion springs in tanks, armour-piercing projectiles, and other ordnance material. Several comprehensive investigations[2, 3] were also conducted, consisting of both laboratory and field service tests of a variety of automobile, tractor, and ordnance parts, so that boron could be used, should further conservation of critical alloys beyond the National Emergency Steels become necessary. These programmes showed that leaner alloy steels containing boron could replace many of the more highly alloyed steels.

The war ended before boron steels were put into general use in ordnance applications other than those mentioned, so that experience with their fabrication, heat-treatment, and use was quite limited. After the war, the production of boron steels dropped to less than 0·5% of the total alloy-steel production, but several large companies adopted boron steels for such applications as tractor parts, crankshafts for Diesel locomotives, and hand tools. Most users were content to use the triple alloy steels of low alloy content with which they had had successful experience, rather than to try the unknown boron steels.

The Korean incident again brought about restrictions in critical alloys in the United States, so that even the relatively lean alloy steels of World War II contained too much alloy. The American Iron and Steel Institute and the Society of Automotive Engineers co-operated in a programme to develop still leaner alloy steels using boron to the fullest extent possible. Each boron steel was designed to replace a conventional alloy steel on an equal hardenability basis, with the resulting increased production of boron steels shown in Fig. 2. At the present time this tonnage is about 8% of U.S. alloy-steel production.

MANUFACTURE OF BORON STEELS

General Principles

The manufacture of boron steels is essentially similar to that of the conventional alloy grades. Neither special steelmaking equipment nor complicated changes in customary practice are required to achieve successful results. The prospective maker of these steels, already versed in the art of making the conventional alloy grades, need be concerned only with certain simple precautions to be taken in the final shaping of the heat in the furnace and in making ladle additions. Experience has shown that when these steels are made with the care given to the conventional alloy grades, the results to be expected after the subsequent mill processing are essentially the same as those obtained with conventional alloy steels.

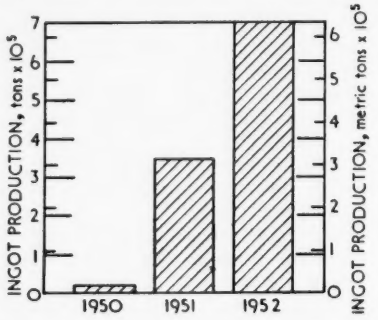


Fig. 2—Approximate production of boron steels in the U.S.A.

Table I

COMPOSITION OF BORON FERRO-ALLOYS USED AS LADLE ADDITION AGENTS

Element	Simple		Complex		
	Ferroboron	Borosil	Carbortam	Grainal 79	Silcaz
Boron, %	12	4	1	0·5	0·6
Silicon, %	4	40	3	5	40
Titanium, %	...	...	16	20	10
Aluminium, %	...	...	1	13	7
Manganese, %	...	...	...	8	...
Zirconium, %	...	...	...	4	4
Calcium, %	...	...	...	...	10
Iron, %	← Balance →				

The most important fact to be considered in the production of boron steel is that satisfactory and reproducible hardenability can be obtained over a wide range of boron content, as shown in Fig. 3. Below about 0·0005% of boron the response is doubtful, and so this is considered to be the lower limit of the desirable boron range, bearing in mind that the limit of accuracy of the boron analysis is ± 0·0003%. The greatest amount of boron that can be used is not defined by the effect of this element on hardenability, but rather by the fact that steels containing more than about 0·007% of boron become 'hot-short' and may crack severely during reduction by rolling or forging. The boron content at which the steel becomes hot-short is well above the amount normally added and is, therefore, of no real concern to the steelmaker.

Boron combines very actively with either oxygen or nitrogen. In fact, one opinion is that its primary function is to tie up the last traces of nitrogen in the steel. In any case, most of the nitrogen must be removed, but this cannot be accomplished by large additions of boron because of the possibility of recovering more than 0·007% of boron, which produces hot-shortness. The major requirement in producing a steel that will contain at least 0·0005% and not more than 0·007% of boron is to render most of the oxygen and nitrogen inactive by means of aluminium, titanium, or other suitable deoxidizers and nitride-formers. Also, to obtain reasonably consistent recovery of boron, it is essential to avoid practices that will cause wide fluctuations in the oxygen and nitrogen contents of the steel before ladle deoxidation, as for example, an inactive bath during the refining period, over-oxidation resulting from excessive bath temperature, an ineffective furnace block, a slow tap, or the early appearance of slag after tap. If these factors are under reasonable control, the production of satisfactory boron steel depends only on the types, amounts, and sequence of the ladle additions.

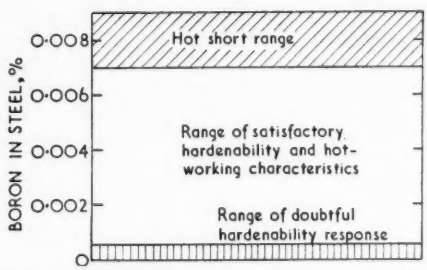


Fig. 3—Permissible range of boron content

In regard to the choice of the boron agent to be added in the ladle, we have found that substantially equal and satisfactory results are obtained from either the simple ferro-alloys of boron (as combined with iron or silicon) or the more complex ferro-alloys, which variously include one or more of the deoxidizers and nitride-formers such as aluminium and titanium. Table I shows the chemical compositions of several boron addition agents, representing both the 'simple' and 'complex' types. The first of these simple ferro-alloys, Ferroboron, has the nominal contents of 12% of boron and 4% of silicon without appreciable amounts of other elements except iron. Borosil is likewise one of the simple ferro-alloys, but differs from Ferroboron in its lower boron and higher silicon contents. In using these simple ferro-alloys, deoxidizers and nitride-formers (sometimes called protective additions) must be added in the ladle before the boron is introduced. The manufacture of boron steels in the United States is based on the use of fine-grain deoxidation practice which, for conventional alloy grades, involves the use of aluminium in the ladle in amounts ranging from 1 to 2½ lb. of aluminium per ton, depending on the plant operating conditions and the ordered carbon level of the steel. In making boron steels, however, this aluminium addition is usually increased by about ½ lb./ton, not only to ensure consistent grain size and the usual deoxidation of the steel, but also to provide sufficient aluminium to assist in rendering the nitrogen inactive. The aluminium is followed by ferrotitanium or similar material as the primary nitride-former. As an example of the use of these simple boron alloys, Table II shows

Table II

TYPICAL SEQUENCE OF LADLE ADDITIONS WITH SIMPLE AND COMPLEX FERRO-ALLOYS

Sequence	Simple Ferro-alloy	Complex Ferro-alloy
1	Coal	Coal
2	Aluminium	Aluminium
3	50% ferrosilicon	50% ferrosilicon
4	75% ferromanganese	75% ferromanganese
5	Aluminium	Aluminium
6	17% ferrotitanium	Grainal 79
7	Borosil	...

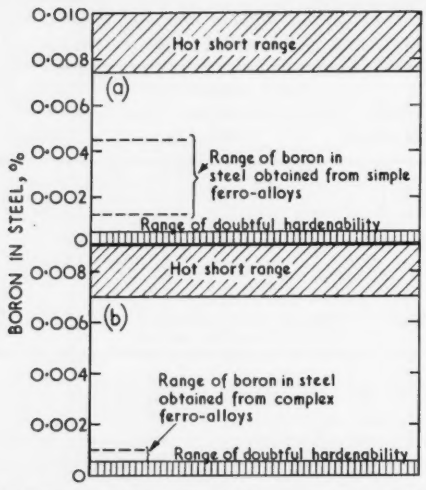


Fig. 4—Ranges of boron added and recovered with (*a*) simple ferro-alloys and (*b*) complex ferro-alloys

a typical sequence of ladle deoxidation, using Borosil. Both the aluminium and the ferrotitanium are added in the ladle before the Borosil addition is made.

Of the complex boron ferro-alloys in Table I, Carbortam has boron and silicon contents lower than those of Borosil, but its titanium content is about 16%, and it contains about 1% of aluminium. Grainal has a lower boron content than Carbortam and their silicon and titanium contents are of the same order, but the percentage of aluminium (13%) in Grainal is quite substantial compared with the much smaller amount found in Carbortam. Grainal also contains zirconium and manganese in appreciable amounts. The boron content of Silcaz is fairly low, as in Grainal, but appreciable amounts of other active elements are present, including nearly 40% of silicon, 10% of titanium, 7% of aluminium, 4% of zirconium, and 10% of calcium.

In contrast to the simple boron ferro-alloys, the complex types carry their own deoxidizers and nitride-formers for the protection of the boron. In the amounts of ferro-alloy used, sufficient nitride-formers are provided. It has been found, however, that additional aluminium, up to ½ lb./ton above that required for conventional alloy steels, is sometimes needed for further protection. The aluminium in the ferro-alloy should be considered in estimating the amount of extra aluminium required. As an example of the use of one of these complex ferro-alloys,

Table IV

TYPICAL LADLE DEOXIDATION OF BASIC ELECTRIC HEATS WITH SIMPLE AND COMPLEX FERRO-ALLOYS

Sequence	Material	Amount*	
		lb./ton	kg./metric ton
Simple Ferro-alloy			
1	Graphite	a.r.	a.r.
2	Aluminium	1	½
3	50% ferrosilicon	a.r.	a.r.
4	Aluminium	1	½
5	17% ferrotitanium	4	2
6	Borosil	2½	1¼
Complex Ferro-alloy			
1	Graphite	a.r.	a.r.
2	Aluminium	1	½
3	50% ferrosilicon	a.r.	a.r.
4	Aluminium	1	½
5	Grainal 79	6	3

* a.r. = as required

Table III

TYPICAL LADLE DEOXIDATION OF OPEN-HEARTH HEATS WITH SIMPLE AND COMPLEX FERRO-ALLOYS

Sequence	Material	Amount*			
		0·20% C		0·40% C	
		lb./ton	kg./metric ton	lb./ton	kg./metric ton
Simple Ferro-alloy					
1	Graphite	a.r.	a.r.	a.r.	a.r.
2	Aluminium	1¼	⅝	1	½
3	50% ferrosilicon	a.r.	a.r.	a.r.	a.r.
4	75% ferromanganese	a.r.	a.r.	a.r.	a.r.
5	Aluminium	1¼	⅝	1	½
6	17% ferrotitanium	2½	1¼	2	1
7	Borosil	2½	1¼	2½	1¼
Complex Ferro-alloy					
1	Graphite	a.r.	a.r.	a.r.	a.r.
2	Aluminium	1¼	⅝	1	½
3	50% ferrosilicon	a.r.	a.r.	a.r.	a.r.
4	75% ferromanganese	a.r.	a.r.	a.r.	a.r.
5	Aluminium	1¼	⅝	1	½
6	Grainal 79	4	2	4	2

* a.r. = as required

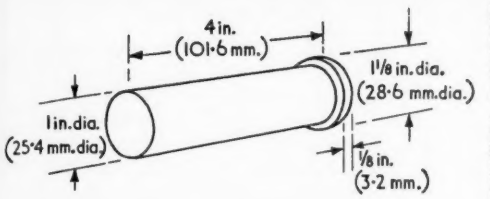


Fig. 5—End-quench hardenability test specimen

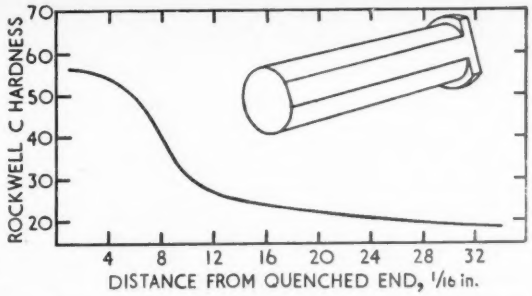


Fig. 6—Typical end-quench hardenability curve obtained from hardness survey of test specimen

Table II shows a typical ladle addition sequence, using the addition agent Grainal. As in the case of the simple ferro-alloy, aluminium is added before the Grainal. The ferro-titanium, however, is omitted.

Since satisfactory hardenability response and ductility at high temperatures are regularly obtained with boron contents of 0·0005–0·007%, the amount of boron ferro-alloy added is not critical. Even after adequate additions of aluminium and ferro-titanium, recovery may vary from 50 to 90%, depending upon carbon content, temperature, and many other factors, but no particular hardship results. When the inexpensive simple boron ferro-alloys are used, the quantity added is calculated to contain the equivalent of 0·0025–0·005% of boron at 100% recovery. As shown in Fig. 4*a*, the resulting boron content in the steel may vary from about 0·0012 to 0·0045%, safely within the required range.

In the case of the more expensive complex boron ferro-alloys, less boron is added (usually 0·001–0·0015% in open-hearth steels) but, as shown in Fig. 4*b*, the recovery is satisfactory, ranging from 0·0005% to about 0·0012%.

Typical Basic Open-Hearth Heat

Up to ladle deoxidation, the production of a boron steel does not differ from that of any conventional alloy steel. Alloy heats are, however, usually 'blocked' 15–20 min. before tap with 0·10% of silicon as 50% ferro-silicon or silico-manganese. Most of the necessary ferromanganese and all of the ferro-chromium are added in the furnace immediately after the ferro-silicon. As shown in Table III, the first addition to the ladle is graphite (or perhaps anthracite coal), if needed to meet the ordered carbon analysis, followed by half the total aluminium required, which is about ½ lb./ton greater than the amount needed for equivalent conventional alloy grades; the first addition of aluminium is thus about 1¼ lb./ton for a 0·20% C heat, or 1 lb./ton for a 0·40% C heat. The practice of splitting the aluminium into two separate additions is customary in some plants to ensure consistent grain-size control in conventional steels. The first aluminium addition is followed by sufficient 50% ferro-silicon to meet the silicon requirement after taking into account the silicon obtained from the block and from the subsequent boron ferro-alloy. A ferro-manganese addition is next made to provide any additional manganese adjustment needed, followed by the second half of the aluminium.

If the heat were being made as a conventional alloy steel without boron, the last aluminium addition, in an amount somewhat less than shown, would have completed the deoxidation. In making a boron heat with a simple boron ferro-alloy, however, the only further requirement, before adding the boron, is to denitrogenize the steel further with a nitride-former, such as 17% ferro-titanium in the amount of 2½ lb./ton of steel for the low-carbon grades and 2 lb./ton for the higher-carbon grades. If desired, an equivalent amount of a zirconium ferro-alloy could be used instead of the ferro-titanium. Ferroboron or Borosil follows the titanium in amounts up to 0·005% of boron calculated at 100% recovery. In the practice shown, the 2½ lb. of Borosil per ton would supply about 0·004% of boron. About 0·0025% of boron would be found in the steel by spectrographic analysis.

When a complex boron ferro-alloy is used in open-hearth practice, an example of which is shown in Table III, the only essential differences in ladle deoxidation are the omission of the ferro-titanium and the larger amount of boron ferro-alloy required because of its lower boron content. With 4 lb. of Grainal per ton added to the ladle as the last addition, about 0·001% of boron is added to the steel, of which about 0·0007% will be recovered.

Boron steels are also made in the basic electric furnace. In this method of manufacture, the same practice used in making conventional alloy grades of steel is again followed up to the ladle deoxidation. We normally use a 100% scrap charge, double-slag practice in which carbidic slags are specified as the second slag for steels above 0·12% of carbon and lime-alumina or lime-silica slags are specified for heats with 0·12% of carbon or less.

As shown in Table IV, the deoxidation of an electric-furnace heat does not differ greatly from that of an open-hearth heat, except that some recognition is given to the fact that electric-furnace steels contain more nitrogen. Graphite or coal is the first ladle addition when recarburization is necessary, and is followed by 1 lb. of aluminium per ton as the first half of the total addition. The amount of aluminium added in the ladle is the same for low- or high-carbon heats, since the lower-carbon heats (under about 0·12% of carbon) are given prior deoxidation in the furnace by means of 1 lb. of aluminium added to the bath just before tapping. The first ladle aluminium is followed by the silicon additions, frequently as calcium silicon, and then the final 1 lb. of aluminium per ton. Again, except for the fact that the total

Table V

DETERMINATION OF CALCULATED IDEAL DIAMETER FOR HEAT ANALYSIS (EXCLUDING BORON)

	Ladle Analysis	Multiplying Factor
C, %	0·44	0·242
Mn, %	0·75	3·500
Si, %	0·26	1·182
Ni, %	0·21	1·077
Cr, %	0·29	1·626
Mo, %	0·01	1·030

$D_i(c) = 0{\cdot}242 \times 3{\cdot}500 \times 1{\cdot}182 \times 1{\cdot}077 \times 1{\cdot}626 \times 1{\cdot}030$
$= 1{\cdot}805$ in.

aluminium has been increased by about $\frac{3}{4}$ lb./ton, the deoxidation practice up to this point is the same as for the conventional alloy grades.

If a simple boron ferro-alloy is to be used as the boron addition agent, a ferro-titanium or other nitride-forming addition must be made. In the electric-furnace process, however, the amount of titanium is about twice that used in the open-hearth, the ferro-alloy being added to the extent of 4 lb. of 17% ferro-titanium per ton or 2 lb. of the 40% grade per ton. It may also be necessary to increase the boron addition slightly above that used in the open-hearth, but, in the practice shown, $2\frac{1}{2}$ lb. of Borosil per ton is used, as in the open-hearth practice.

When one of the complex boron ferro-alloys, *e.g.*, Grainal, is used as a ladle addition (*see* Table IV), it is customary, as in the open-hearth, to omit the titanium because the necessary nitride-formers are included in the material. The amount of boron agent, however, is increased by 50% over that needed in the open-hearth.

The teeming and subsequent processing of a boron heat through the soaking pits and mills is similar to that of the conventional alloy steels. Since boron steels are less susceptible to flaking than conventional alloy steels of equivalent hardenability, the controlled cooling cycle may be reduced. As an example, an S.A.E.4617 steel, containing about 0·17% C, 1·85% Ni, and 0·25% Mo, would be given a retarded cool from above the critical temperature in covered insulated railroad cars or pits, whereas the equivalent boron steel, A.I.S.I. 94*B*17, containing about 0·17% C, 0·45% Ni, 0·40% Cr, and 0·12% Mo, would require only air-cooling. It has also been reported that boron steels suffer less from such things as sticky adherent scale by virtue of their lower alloy content and that they tend to be softer in the unhardened condition.

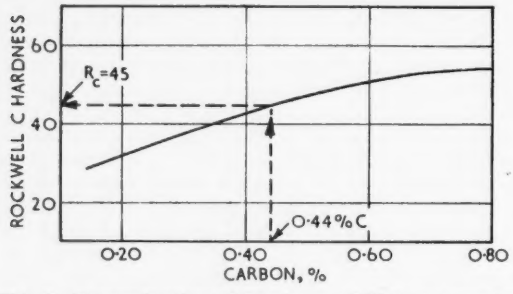


Fig. 7—Determination of hardness of 50% martensite for a 0·44% C steel

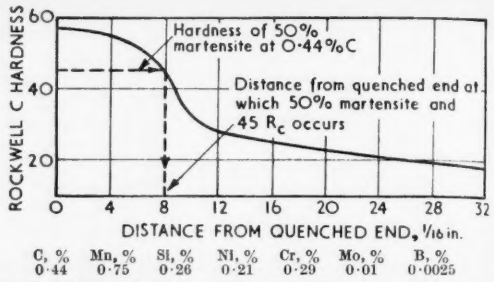


Fig. 8—Determination of distance from quenched end at which 50% martensite recurs (grade *TS50B46*)

The only reliable test of the success of the boron treatment is the end-quench hardenability test. A short section of the rolled product is obtained at the shears after primary mill rolling and is subsequently forged into a $1\frac{1}{4}$-in. round bar. These tests are ordinarily obtained from product representing the middle of the middle ingot and the top of the last ingot of the heat. The forged rounds are normalized and machined to the standard end-quench bar of dimensions shown in Fig. 5.

The end-quench test bar is heated to the prescribed temperature, held for 20 min., and then placed in a fixture. A water jet is turned on immediately, impinging on one end of the bar. Since the opposite end cools in air, all cooling rates from water-quenching to air-cooling are obtained along the length of the bar. Consequently, the hardening characteristics of the heat of steel are determined by making a hardness survey along the length of the bar. A typical hardness curve obtained from the test is shown in Fig. 6. The complete details for making this test are found in the S.A.E. Handbook or in the A.S.T.M. Specification A–255.

To determine the increase in hardenability effected by an addition of boron to a steel, an indirect approach is used, since the amount of boron present is not necessarily a measure of the hardenability increase. This calculation is begun by multiplying together Grossmann's factors for each element except boron, obtaining the ideal diameter at 50% martensite. This computation is shown in Table V for a 0·44% C *TS50B46* heat, where a calculated ideal diameter $D_i(c)$ of 1·805 in. is obtained. The actual ideal diameter $D_i(a)$ of the steel with boron is then determined from the end-quench test results. As shown in Fig. 7, the Rockwell *C* hardness of 50% martensite

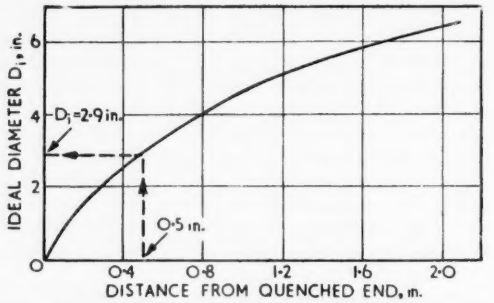


Fig. 9—Conversion of Jominy distance to ideal diameter

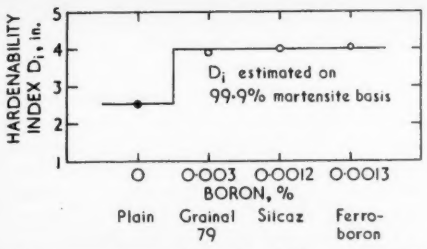

HARDENABILITY INDEX D_i, in.
D_i estimated on 99·9% martensite basis
0 0·0003 0·0012 0·0013
BORON, %
Plain Grainal 79 Silcaz Ferro-boron

Fig. 10—Comparison of hardenability of S.A.E. 4150 and 41*B*50 steels

at the given carbon content (0·44% in the example) is determined from the appropriate Hodge–Orehoski curve,[4] providing a value in the example of 45 R_c. Translating this value to the end-quench hardenability curve shown in Fig. 8, for the sample heat, 50% martensite is found at a distance of $\frac{1}{2}$ in. Transferred to a second Hodge–Orehoski curve,[5] shown in Fig. 9, the actual ideal diameter $D_i(a)$ for the sample heat is read as 2·9 in.

Once the theoretical hardenability of a heat without boron (expressed as a calculated ideal diameter) and the actual hardenability of a heat with the same analysis but with boron (also expressed as an ideal diameter) have been determined, the boron factor can be obtained by dividing the actual by the calculated value, the quotient indicating the effectiveness of the boron in increasing hardenability.

$$\text{Boron factor } F_b = \frac{D_i(a)}{D_i(c)} = \frac{2{\cdot}9}{1{\cdot}80} = 1{\cdot}61$$

where $D_i(a)$ = actual ideal diameter of boron steel as determined from Jominy values

$D_i(c)$ = calculated ideal diameter of steel with similar analysis without boron, calculated from actual analysis excluding boron.

From our own experience on a large number of heats, we have determined that the average factor for boron for this grade of steel is 1·42. The indicated result of 1·61 confirms that the expected level of hardenability has been attained.

EFFECT OF TYPE OF BORON ADDITION ON PROPERTIES OF BORON STEELS

Several investigations have been conducted to determine the relative merits of the various ferro-

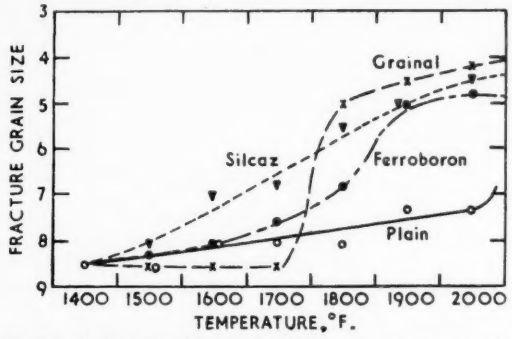


Fig. 11—Influence of boron addition agent on austenite grain coarsening in S.A.E. 4150 steel

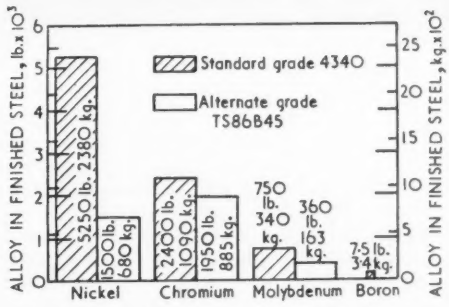


Fig. 12—Conservation of alloys by use of boron; amounts added to 150-ton open-hearth heats of grades 4340 and *TS*86*B*45

alloys used in boron steelmaking in regard to their effects on the properties of the steel. One of the most comprehensive studies was that conducted by the U.S. Steel Corporation,[6] in which four 150-ton open-hearth heats of S.A.E. 1046 (0·45% C, 0·80% Mn), four of S.A.E. 1321 (0·20% C, 2·00% Mn), and four 70-ton electric-furnace heats of S.A.E. 4150 (0·50% C, 0·85% Mn, 1·00% Cr, 0·20% Mo) were made and tested completely. One heat of each grade was made with no boron addition, and a different type of boron ferro-alloy was added to each of the others by the standard production practice.

Figure 10 shows the hardenability effect in S.A.E. 4150 in terms of the ideal diameter (D_i) as determined from the end-quench hardenability results for the three ferro-alloys Grainal 79, Silcaz 3, and Ferroboron. The hardenability was the same for all three heats, and all are substantially higher than the heat containing no addition. Similar results were obtained for the S.A.E. 1046 and 1321 series.

Another important characteristic of alloy steels is grain-growth behaviour. Figure 11 shows the austenitic grain size of S.A.E. 4150 at various temperatures. It is evident that all the S.A.E. 4150 heats containing boron coarsened at a lower temperature than the heat containing no boron. Although the Ferroboron heat seemed to be slightly superior to the other two boron heats in this series, this behaviour was not consistent in the other two grades. As a result of these findings, increased aluminium is now used by our Corporation for all boron heats, regardless of the ferro-alloy used, so that steels having the fine-grain characteristics desired by users in the United States can be produced.

Many other tests, such as tensile, impact, cleanliness, macro-etch, and fatigue, were conducted besides those on hardenability and grain size. The conclusion of this investigation was that the properties of boron

Table VI

SAVINGS OF NICKEL AND MOLYBDENUM WITH BORON STEELS

Analysis	Grade 4620	Grade *TS*94*B*17
C, %	0·17	0·17
Ni, %	1·83	0·45
Cr, %	...	0·40
Mo, %	0·25	0·12
Savings Using Boron Steel	**lb./ton**	**kg./metric ton**
Nickel	27·6	13·8
Molybdenum	2·6	1·3

steels are independent of the type of boron ferro-alloy used.

CONSERVATION OF ALLOYS EFFECTED BY USE OF BORON

In the United States, the primary purpose in using boron has been to conserve the critical alloys—nickel, molybdenum, and chromium. The following examples from our experience will indicate the extent to which this has been effected.

Figure 12 shows the amount of alloys saved in a 150-ton (136-metric ton) open-hearth heat where a grade of steel known as S.A.E. *TS86B45* (0·45% C, 0·80% Mn, 0·55% Ni, 0·65% Cr, 0·12% Mo) is now being made in place of S.A.E. 4340 (0·40% C, 0·70% Mn, 1·80% Ni, 0·80% Cr, 0·25% Mo), which was formerly used. The total saving in critical alloys is 4590 lb. (2081 kg.).

Substantial savings can be effected in most of the conventional alloy steels. Table VI shows a typical alloy steel together with the recommended boron steel and the amounts of critical alloys that can be saved by the use of boron. It is possible to save 27·6 lb. of nickel per ton and 2·6 lb. of molybdenum per ton. In 1951, the use of boron steels in the United States was responsible directly or indirectly for the saving of about 12,000,000 lb. (5,450,000 kg.) of nickel, which was thereby made available for more essential applications.

SUMMARY

The development of the manufacture and use of boron steels in the United States is merely another step in the trend towards better utilization of available alloying elements. Although there are some peculiarities associated with the manufacture of boron steels, they can be made with only minor adjustments in steelmaking practice. When boron steels are properly applied, a very small quantity of boron can replace sizeable quantities of critical alloys.

References

1. " Report on Special Alloy Addition Agents," Contributions to the Metallurgy of Steel No. 9: 1942, New York, American Iron and Steel Institute.
2. " Report on Investigation of Boron-Treated Steels: Ordnance Project," Spec. Publ. No. 10: 1946, New York, Society of Automotive Engineers.
3. " Report on Investigation of Boron-Treated Steels: Caterpillar Project," Spec. Publ. No. 18: New York, Society of Automotive Engineers.
4. J. M. HODGE and M. A. OREHOSKI: *Metals Technology*, 1945, vol. 12, *Tech. Publ. No.* 1800.
5. J. M. HODGE and M. A. OREHOSKI: *Ibid.*, 1946, vol. 13, *Tech. Publ. No.* 1994.
6. R. A. GRANGE, W. B. SEENS, W. S. HOLT, and T. M. GARVEY: *Amer. Soc. Metals, Preprint No.* 5, 1949.

DISCUSSION

Mr. **Rohl,** after reading his paper, said: Any progress made in the utilization of these steels must be a joint effort from the application side and the manufacturing side. Neither producer nor user alone can achieve success, and it is only by a combination of many producers and many users contributing their practices and results that success will be achieved in the utilization of boron steel.

I asked some of our people to say, on the basis of their experience, to what extent the boron ranges they use would replace other elements. This information (*see* Table *A*), whilst it may be subject to some variation, gives some conception of the extent to which the boron addition can replace other elements, on the basis of physical properties.

The **Chairman:** The paper was excellent and the manner of its presentation delightful. As I listened to it, my mind went back to the time when vast quantities of Sheffield water were shipped to the United States in barrels, because it was believed that there were ingredients in Sheffield water that increased the hardenability of steels. A great deal of progress has been made since those days, and Mr. Rohl now has two important advantages: his advocacy of the use of boron is backed up by a vast amount of scientific knowledge and experiment, and the quantity of boron required compares very favourably with the quantity of Sheffield water that must have been necessary.

Dr. **H. H. Burton** (English Steel Corporation, Ltd.): As I think that there is some misapprehension amongst our friends in the U.S.A. on the subject of the British attitude towards boron steels in general, I should like to assure them, through Mr. Rohl, that there is no lack of interest in these steels in the U.K. We welcome papers on the subject, particularly when, like this paper, they are backed by a wealth of first-hand experience. Our reluctance to accept boron steels so wholeheartedly as they have been accepted in the U.S.A. is due largely to differences between our respective ideas of specifications, and also to some doubt about the reproducibility of the results, due to variables in steelmaking when making boron steels that have hitherto not been fully understood. Mr. Rohl's paper has done much to clear up some of those difficulties.

Mr. Rohl referred to the upper limit where the addition of 0·007% or more of boron produces hot-shortness. Has much work been done on boron steels in the cast condition? If so, is it established what form of breakdown occurs when the steel exhibits this hot-shortness, and is it intergranular? An investigation on those lines might yield very interesting results, because, once what is happening in the cast condition is known, we have gone a long way towards solving the problem of what happens in the wrought condition.

What degree of uniformity is expected in a particular cast from the first to the last ingot? In many fine-grained steels, however carefully they are made, there

Table A

AMOUNTS OF INDIVIDUAL ALLOYING ELEMENTS HAVING HARDENABILITY EFFECT EQUIVALENT TO BORON

Base Composition		Amount of Individual Alloy Needed to be Added to Base Composition to be Equivalent to Boron			
C, %	Mn, %	Mn, %	Ni, %	Cr, %	Mo, %
0·20	0·75	0·85	2·40	0·45	0·35
0·40	0·75	0·65	1·90	0·35	0·25
0·60	0·75	0·45	1·20	0·20	0·15
0·80	0·75	0·15	0·40	0·07	0·05

are often mixed grain sizes. In making boron steels, are mixed hardenabilities obtained?

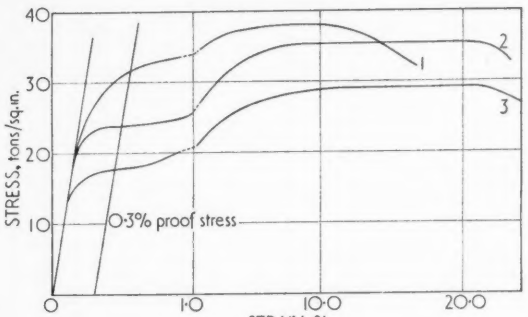
What is the result of melting boron steels from boron-bearing scrap? Is there any cumulative effect, and must that be allowed for in making boron additions? If the effect is at all cumulative, a point might be reached where we would get hot-shortness without suspecting it.

Mr. Rohl showed in his Table VI some results of the substitution of *TS*94*B*17 for 4620. Alloy 4620 has hitherto been very widely used in the U.S.A., particularly for bearings; how far has that substitution gone? Is a great deal of 4620 steel not still used, because of certain virtues it possesses that *TS*94*B*17 does not possess?

Mr. **Rohl**: We do not have much first-hand knowledge about the breakdown of the steel. I suspect that it may be intergranular. With over 0·007% B rolling conditions are very bad, whereas the steel might normally roll very well. We have considerable trouble with surface breakage and deep cracks, which are characteristic of hot-short steel, but I cannot answer the question of how it occurs so far as grain structure is concerned.

It has been our experience that we do have very good uniformity. We originally took tests from the bottom of the first ingot, the middle of the middle ingot, and the top of the last ingot of the heat. Experience taught us that we could abandon the test on the first ingot, because there was very little variation. In our plant we always carry out a test on the last ingot, whether it is boron steel or a standard alloy steel. To maintain the grain size, there are instances where in particular heats ½ or ¼ lb. of pebbled aluminium may be added throughout the ingot to ensure that the grain size is maintained and coarsening is avoided. In some cases pebbles of aluminium are therefore added in the last three or four ingots of a 42-ingot heat. Our heats range from 100 to 275 tons/cast, and, depending on the observations of the metallurgist who follows each of those heats, an addition of that kind may be made. Apart from that, the uniformity is exceedingly good.

It has not been our experience that we get any boron recovery from the scrap. In the several hundred heats which we have made there has not been any evidence of residual boron in boron heats or in any other heats where boron-containing scrap is used.



Steel	0·3% Proof Stress, tons/sq. in.	Max. Stress, tons/sq. in.
1. Fortiweld	31·5	38·05
2. Steel to B.S.S. 968 ..	23·9	35·2
3. Mild steel	17·7	29·2

All in form of 1-in. thick as-rolled plate

Fig. A—Stress/strain curves for mild and high-tensile steels

The 4620 steel has had a very wide use for bearings, but it is just not available to-day, and the substitutes that have been made are those which have been recommended and adopted on a very broad scale, because within the U.S.A., owing to the restrictions that have been in force, it has not been possible to use the 4620 steel for that purpose. The substitutes used for 4620 in bearing manufacture have been either *TS*4720 or *TS*8620, neither of which contains boron. The reason is that the primary requirement is hardenability of the carburized case, and control of distortion has been a problem. However, thousands of tons of *TS*94B17 have replaced 4620 successfully for automobile and truck gears.

Dr. **L. Reeve** (Appleby-Frodingham Steel Co.): In the United Steel Companies interest in boron steels has been very largely, though not entirely, confined to a structural molybdenum–boron steel used in the as-rolled or normalized condition and not fully quenched and tempered. Our method of making these boron steels is directed entirely to obtaining the best properties of the finished steel essentially in the as-rolled condition.

In giving the boron ranges used in making these steels, Mr. Rohl did not specify whether he was speaking of soluble or insoluble boron. We have made these boron steels using both simple and complex deoxidizers. There is a difference, in that on the whole the complex deoxidizers tend to give almost no boron in the so-called insoluble form, if the steel has been correctly made; it is almost all present as soluble boron. When we use the simple deoxidizers, which we prefer for the boron steels in the as-rolled condition, the recovery of soluble boron is not so high ; there is more insoluble boron, which seems to play no part in the final properties of the finished steel. Does Mr. Rohl use more of the complex deoxidizers than of the simple ones, and has he any further information on this question of the amount of boron in the soluble form?

There are a number of points of major difference between our practice and that described by Mr. Rohl. His normal practice is to block the heat in the final stages and tap 20 min. later. Under the conditions at Appleby-Frodingham, using a highly phosphoric charge with a high-phosphorus slag, we cannot block the bath, and we have to tap in the fully oxidized condition, so that the degree of oxidation of the bath before tapping is important. We add all our deoxidizers and additions in general to the ladle. The general procedure described in the paper is obviously based on the same background idea, *i.e.*, to put all the major deoxidizers in before adding the boron.

We do the same: we add the de-oxidizers, including aluminium, fairly early; finally, but before the ladle is two-thirds full, the boron is added, usually as ferro-boron. Under those conditions, the variation of the heat from beginning to end of the cast has been studied by taking a series of pit samples for every individual ingot ; the difference between the first and the last ingot is within the experimental error of the boron determination. Reaction with slag is avoided in our practice, since there is normally none coming over from the tilting furnaces.

Some of the mechanical properties of the Mo–B steel which we are making are illustrated in Fig. *A*, which shows the stress/strain diagrams for this steel, a 1½% Mn steel, and a mild steel. The scale from 0 to 1·0% strain is deliberately magnified ten times as compared with the higher strain values. The Mo–B steel has a yield point (0·3% proof stress) of 31·5 tons/sq. in. and a maximum stress of just over 38 tons/sq. in. A similar steel without boron would have had a yield point at best of 18 tons/sq. in., and, more probably, in

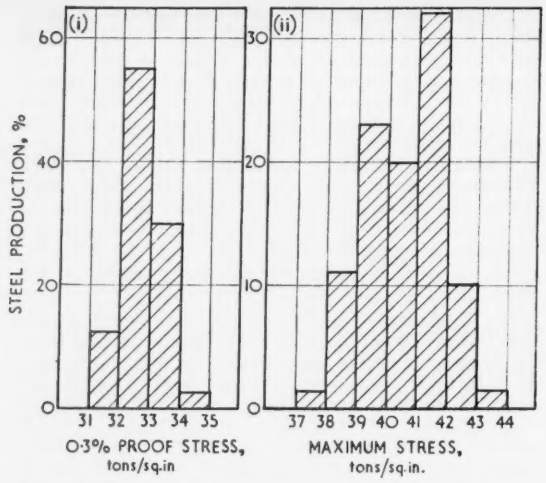


Fig. B—Distribution of 0·3% proof stress and maximum stress for production tonnage of Mo–B steel

view of the low carbon content (usually below 0·12%), of 15–16 tons/sq. in. The effect of the 0·0015–0·0035% of soluble boron has been to almost double the yield point of what is otherwise a very-low-carbon ½% Mo steel.

The 1½% Mn structural steel has a yield point of 23·9 tons/sq. in., and a maximum stress of 35·2 tons, and the mild steel has a yield point of 17·7 tons/sq. in. and a maximum stress of 29·2 tons/sq. in.

Figure *B* is a histogram showing the distribution of the 0·3% proof stress and maximum stress for a substantial production tonnage of Mo–B steel in the form of 6 in. × 3 in. channels in the as-rolled condition. About 85% of the steel has a 0·3% proof stress between 32 and 34 tons/sq. in., and about 75% of it had a maximum stress in the range 39–42 tons/sq. in.

Figure *C*(i) is a histogram showing the soluble boron in these casts. Over 90% of the steel had a soluble boron content between 0·0020 and 0·0030%. Figure *C*(ii) shows the distribution of the Izod impact tests taken on these steels, again in the as-rolled condition. Over 90% of the steel had impacts exceeding 50 ft. lb.

For material in the as-rolled condition the degree of variability shown in these histograms may be regarded as satisfactory. The variability has not been enough to cause any serious production difficulties and we have been able to meet specifications without much trouble.

Mr. **Rohl**: In speaking of boron I had in mind the total boron, which is the basis on which our determinations are made. Whether or not the complex form is used will vary with different companies and is largely related to economy. We originally used ferro-boron in the production of armour plate during the war period. Subsequently, as the field of applications was extended, our plants began to use so-called ' complex ' ferro-alloys of the Grainal type. Later, with more experience, they adopted the use of simple ferro-alloys of the Borosil type. However, other large producers feel that there is greater safety in using the complex, even though it may be more expensive. I cannot say definitely, therefore, that there is an advantage in one or the other; it is a question of economy. We watch the performance constantly. We have eleven plants, four of which produce these boron steels, and we have monthly meetings to determine what their results are. They are at liberty to use whichever alloy they choose. Being largely based on what they consider to be the most economic, it may be subject to change. We use the simple, whereas others use the complex quite satisfactorily.

I was very interested in Figs. *A–C*. In the U.S.A. we are not at liberty to do much about making these steels, because we are not allowed to use so much molybdenum,but many people are interested in the characteristics of this steel, and, should there be a liberalization of the position in regard to molybdenum, they might go further than they have in its utilization. We have done some work on steels somewhat comparable to these, although in our case we resorted to quenched and tempered plate. The weldability seemed good, and there were indications that a minimum of pre- and post-heating was necessary.

Monsieur **A. Kohn** (I.R.S.I.D.): As early as 1947, French metallurgists were interested in boron steels, and at that time some casts were melted in high-frequency furnaces; two industrial casts were also produced in a 10-ton basic electric-arc furnace, one a plain carbon steel and the other a low-alloy steel. Boron was incorporated as ferro-boron after deoxidation with aluminium. The effect of boron on the hardenability of the steels was evident, but its action was effective only when the boron addition was three times higher than the maximum content claimed by the American workers. The determination of the boron content by Rudoph and Flickinger's method[1] showed that a large part of the boron had been fixed in a form insoluble in sulphuric acid; this seemed to be a boron nitride.

These results led us to think that nitrogen present in the steel might fix the boron in an ineffective form, and that more than 0·001% of boron soluble in sulphuric acid was necessary to have some effect on the properties of the metal, as Dr. Reeve has said.

This troublesome action of nitrogen was clearly supported by Mr. Speight's work.[2] The purpose of these first researches was to make more precise the effect of boron upon the general properties of steels. Further investigations were carried out in France in order to manufacture boron steels correctly and to com-

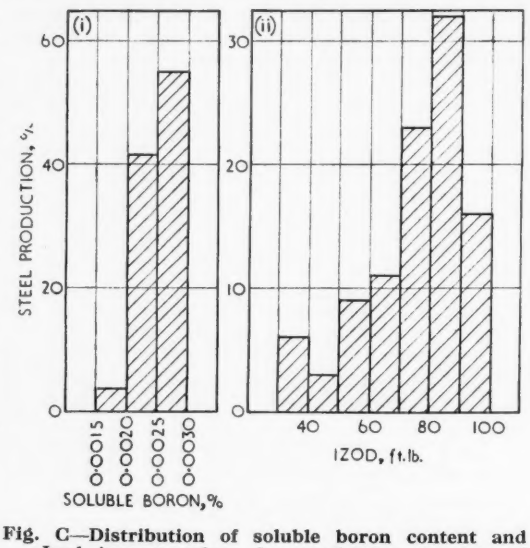


Fig. C—Distribution of soluble boron content and Izod impact values for production tonnage of Mo–B steel

Table B

CHEMICAL COMPOSITION OF THE BORON STEELS

Boron = 0·0015–0·0020%

	Equivalent A.I.S.I. grade	C, %	Mn, %	Ni, %	Cr, %	Mo, %
A	94*B*17	0·17	0·80	0·65	0·58	0·18
B	80*B*15	0·19	0·79	0·24	0·30	0·24
C	81*B*35	0·34	0·85	0·32	0·44	0·17
D	81*B*35	0·36	0·72	0·27	0·50	0·12
E	86*B*45	0·44	0·98	0·54	0·60	0·12

pare them with higher-alloy steels. Five casts were made by the Anciens Etablissements Cail, Les Aciéries d'Ugine, and la Société des Forges et Ateliers du Creusot, and tested in the laboratories of these companies. The investigations were planned by the French Iron and Steel Research Institute. One of these casts was melted in a 40-ton basic open-hearth furnace and the others in basic electric furnaces. Boron was incorporated during tapping as a complex ferro-alloy (Grainal 79). In two of these casts, boron was added to only one half of the melted steel; half these casts gave boron-bearing ingots and the other half boron-free ingots. It was thus possible to determine the exact effect of boron on the steel under investigation.

Table *B* shows the chemical compositions of these steels. They correspond to low-alloy American steels, as indicated in the first column. The first two steels, *A* and *B*, correspond to carburizing grades. Steels *C* and *D* are very similar in composition, but were made by two different companies. As can be seen, two of these steels have a molybdenum content higher than the A.I.S.I. standard specification content, *i.e.*, 0·17%. Therefore, no ferro-molybdenum has been added and the molybdenum came only from the scrap. This fact indicates that when melting alloy scraps, a common practice in French steelmaking, the molybdenum content may be high. It can then be regarded as unnecessary to specify so low a molybdenum content as in the United States; furthermore, molybdenum gives the metal resistance to temper-brittleness, which cannot be obtained with boron, and improves the boron effect, as Bardgett and Reeve have shown.[3]

At the afternoon session I shall give some further information on the mechanical properties of these steels. For the moment I can say that they have a better hardenability than the boron-free steels of the same grades and as good a hardenability as higher alloy steels. There does not seem to be any difficulty in making correct boron steels, even in the electric furnace, provided that the oxygen and nitrogen present in the melted steel are fixed as stable compounds by convenient elements. This result supports our previous experiments and is in good accord with all the information given by English and American workers.

Mr. **Rohl:** Monsieur Kohn has confirmed our own experience, that the form in which you have the boron determines to a large extent its effectiveness in the hardenability characteristics. It is for that reason that we do not rely on the boron content *per se*, but rather on the hardenability. We make determinations on the basis of the total boron, and we realize that it is necessary to tie up the nitrogen and the oxygen thoroughly. If a condition of oxidation is present we may get suitable boron recoveries, but we shall not necessarily get the hardness that would be obtainable if the nitrogen and oxygen were fully tied up before the boron is added.

Mr. **L. G. E. Morse** (Brymbo Steel Works, Ltd.): My company has made a few casts of boron-containing steel, to be processed into $\frac{5}{8}$-in. dia. hardened and tempered bolts, so that we are more interested in the mechanical properties of the bolts than in their hardenability. With no alloy additions other than the 'tramps' in the scrap we obtained a result of the order of En 16 when suitably hardened and tempered. The additions of boron were made in Grainal form to electric-furnace heats. Putting the Grainal into the stream presented no particular difficulty, and the boron content was about right at the first attempt. The material rolled very well.

Mr. **G. E. Speight** (United Steel Companies, Ltd.): Mr. Rohl said that with the simple alloys he would add about 0·0044% of boron, but with the complex alloys about 0·0012%. That seems curious, because presumably a certain minimum amount of boron is necessary in these steels to give the hardenability; I should have thought that no matter what agent was used it would be necessary to add enough to give the same amount of boron in the steel.

The clue is probably to be found in the fact that these alloys are very expensive; in addition they are of low boron content, so that the physical difficulties of adding the large amount of alloy that would be needed to give, say, 0·002–0·003% of boron in the steel would be great.

Mr. Rohl mentioned that these complex alloys came into use as complex deoxidants, and it was only by accident that they were found to contain boron, which gave this marked effect on hardenability. If that is so, why, when definite additions of boron are required, do

Table C

TYPICAL BORON TREATMENTS OF LOW-ALLOY STEELS

18-lb. H.F. melts

Treatment	Additions, %			Boron, %	
	B	Al	Ti	Acid Soluble	Acid Insoluble
1. B	0·005	—	—	0·0015	0·0024
2. B + Al					
(*a*) High Al	0·003	0·2	—	0·0004	0·0019
(*b*) Low Al	0·006	0·07$_5$	—	0·0023	0·0028
3. B + Al + Ti					
(*a*) High Al, low Ti	0·006	0·3	0·015	0·0046	0·0012
(*b*) Low Al, high Ti	0·003	0·07$_5$	0·08	0·0031	<0·0002

Table D

BASIC OPEN-HEARTH FURNACES

60–75-ton Fixed Furnaces

Bath Additions

Spiegel to block bath
FeMn: (part)

Ladle Additions

FeSi (75%): To specification + 0·05%
FeMn (77%): To specification
Al: 3 lb./ton
B.A.T.S. (1·85%)
(*a*) <0·15% C: 5·8 lb./ton = 0·0048% B
(*b*) 0·30–0·35% C: 4·3 lb./ton = 0·0036% B

Boron Analysis

Total boron: 0·003–0·004%
(*a*) Acid soluble: 0·003–0·0038%
(*b*) Acid insoluble: <0·0002%

(*a*) Boron yield greater with higher carbon heats
(*b*) Boron yield uniform throughout cast; tends to fall away slightly in extra ingots

they continue to use these unsuitable alloys of low boron content? I suggest that the suppliers of these complex alloys should be asked to make an alloy that is more suitable, *i.e.*, one with a higher boron content, so that it would be possible to add, within a reasonable weight of addition, the same amount of boron that is added with Borosil or ferro-boron.

Using the Grainal practice, with 0·0007% of boron, there will be some danger of fading towards the end of the cast, and the later ingots will probably not give the same depth hardenability as the first.

The fact that these steels are not subject to hairline cracks, or not to the same extent as the steels which they replace, is presumably due to the lower alloy content.

The making of boron steels with a controlled yield of effective boron content is relatively easy provided that two points are appreciated:

(i) Boron has a high affinity for both oxygen and nitrogen, and

(ii) The contents of reactive oxygen and nitrogen are in general higher than the desired amount of boron.

Reactions between these elements leading to loss of boron are, therefore, of greater significance than the similar reactions occurring with other deoxidizers and denitridizers. To ensure a high yield of effective boron, the boron addition should be supplemented by additions of other elements having a greater affinity for oxygen and nitrogen, *e.g.*, aluminium and titanium. However, using the above principles, it is possible to devise several methods of adding boron to steel, and Table *C* illustrates some of these applied to manganese- and silicon-killed melts in an 18-lb. H.F. furnace. For example, boron alone may be used in sufficient excess not only to look after the reactive oxygen and nitrogen, but also to provide the desired amount of acid-soluble boron in the steel. In the example quoted, 0·005% of boron was added and, although this yielded 0·0015% of acid-soluble boron, much of this would probably exist as boron oxide and would not be effective in the metallurgical sense.

The presence of deoxidizers, *e.g.*, aluminium, would prevent the formation of the useless boron oxide and increase the yield of effective acid-soluble boron. The two examples quoted in Table *C* illustrate the results from a small amount of boron with a high amount of aluminium, and *vice versa*. These examples are extreme cases and, since aluminium does not prevent some loss of boron as the acid-insoluble compound, a more satisfactory yield of acid-soluble boron (*i.e.*, 0·0015%) would be obtained by using 0·0045% of boron with 0·2% of aluminium.

As already stated, the best yields of acid-soluble boron are obtained when a strong deoxidizer and denitridizer are used with the boron, and Table *C* illustrates the effects of supplementary additions of aluminium and titanium. With a low titanium addition, *i.e.*, 0·015%, some insoluble boron compound still persisted, but this was prevented and the boron was thrown almost entirely into the effective acid-soluble form by an addition of about 0·08% of titanium. Techniques 3*a* and 3*b* have both been used on the commercial scale with completely satisfactory results.

Table *D* summarizes the bath and ladle additions for method 3*b* (Table *C*) applied to 60–75-ton basic open-hearth heats. After blocking the bath with spiegel and adding ferro-manganese, the metal is poured on to ferro-silicon in the ladle, and ferro-manganese, aluminium, and a complex boron alloy containing 1·85% of boron are added during tapping. This complex alloy with its relatively high boron content gives almost complete recovery of the added boron in the desired acid-soluble form.

Table *E* shows the practices that have been adapted to the use of various boron compounds, *e.g.*, ferro-boron,

Table E

TYPICAL BORON-TREATED HEATS

	Fe–B	Complex Boron Alloy (B.A.T.S.)	Borate
Basic Electric-Arc Furnace (12½ tons)			
Additions			
Bath			
Al	—	—	1
Borate (12½)	—	—	4
Ladle			
Al	2½	1	—
FeTi (40%)	5	1	4
FeB (16%)	0·6	—	—
B.A.T.S. (1·85%)	—	5⅓	—
Boron Addition	0·0044%	0·0044%	0·0044%
Analysis			
Boron, %	0·0028–0·0032	0·0032–0·0036‡	0·0048§
(*a*) Acid soluble	0·0030	0·0034	0·0018
(*b*) Acid insoluble	<0·0002	<0·0002	0·0030
Basic H.F. Furnace (12 cwt. and 2 tons)			
Additions			
Bath			
Al	2	1	2
FeTi (40%)	6	—	6
FeB (14%)	0·5	—	—
B.A.T.S. (1·85%)	—	4·8	—
Borax (18·5%)	—	—	1⅓
Ladle			
Al*	1	1	1
Boron Addition	0·0034%	0·0040%	0·011%
Analysis			
Boron, %	0·0031–0·0045†	0·0037	0·0033
(*a*) Acid soluble	0·0030	0·0035	0·0033
(*b*) Acid insoluble	<0·0002	<0·0002	Nil

* Reduce with higher-carbon steels
† Possibility of boron increment from H.F. furnace lining
‡ Improved yield from complex alloy
§ Specification limits on Al addition: 1½ lb. to ladle required

the complex alloy or borate in 12½-ton basic electric-arc furnace heats, together with the comparable practices for commercial basic H.F. furnace heats. In all instances where specifications, etc., permit the use of the supplementary additions of aluminium and titanium, the yield of acid-soluble boron is high and the formation of the acid-insoluble form almost entirely prevented.

Mr. **Rohl**: Mr. Speight's point about tying up oxygen and nitrogen with aluminium and titanium is one which we have learned to be essential.

We have conferred with the ferro-alloy producers on the question of a standard alloy, but we seem to have little influence in the matter of what they choose to produce. Each feels that his own alloy is the most suitable. Borosil is perhaps an outcome of earlier discussions about complex alloys and was designed to be between the ferro-boron straight addition with the higher boron content and the complex. We too would like to see a standard alloy with a higher boron content and the deoxidizers present, but so far we have not been influential in bringing that about.

We have not had much experience of boron recovery from the slag. We have tried to ensure that we get uniformity so far as possible, and we have not gone into that particular phase of it. I believe that some experimental work has been done, but not enough to ensure uniformity of practice, and we have resorted to using either the complex or the simple, largely because of our experience with these. We might gain it from another source, or even from another combination of a ferro-alloy. I am glad to hear his views on that.

Mr. **G. Mayer** (Mond Nickel Co., Ltd.): In making and processing small laboratory chill-cast ingots of boron-treated steels we have not experienced any special difficulties from hot-shortness, but during the examination of a sand-cast test bar of a boron-treated low-alloy steel I found an interdendritic constituent, which was not definitely identified but which was thought to be the iron–boron–carbon eutectic, which is reputed to cause hot-shortness. This steel was treated with only 0·003% of boron, added as the alloy Silcaz.

Figure *D*(i) is a micrograph of the cast steel after partial isothermal transformation at 500° C. to reveal the dendritic structures. The constituent near the centre is the one to which I am referring, and it is obvious that it has formed as a result of segregation. Figure *D*(ii) is another field of the same sample, at a higher magnification. The matrix consists of martensite and bainite, and there is also some pearlitic constituent present. The constituent which is believed to be the iron–boron–carbon eutectic is shown to be duplex in this micrograph.

Has Mr. Rohl encountered such a constituent in steelworks ingots in which the slow cooling rate is likely to promote its formation, and have any difficulties occurred in the early stages of forging and rolling that could be ascribed to its presence?

Mr. **Rohl**: In the early work at our laboratory, Dr. Grange and his associates made many studies of this kind, but I cannot define precisely the type of structure we had. His later work showed that the effective boron, in his view, was in the grain-boundary lines. A paper by Grange and Garvey[4] showed the presence of the soluble (in austenite) boron in the grain-boundary lines; this was the boron, in their view, that was effective in creating the hardenability at which we are aiming.

I cannot definitely identify the structure as being related to the hot-shortness encountered at 0·007% of boron, but this might answer Dr. Burton's question. In the paper by Grange and Garvey, it was shown that the effective boron was in the grain-boundary lines. Even under this condition, the trouble was not encountered until the boron content reached 0·007%, where it is likely that compounds such as nitrides may have caused the trouble.

Mr. **R. J. Brown** (Morris Motors, Ltd.): Mr. Rohl referred to the control of the quality of the steel by hardenability, but obviously related to composition. In the U.K. steels are purchased according to composition and to mechanical properties. I should like some information from Mr. Rohl on the relationship between the user and the steelmaker in the selection and supply of steel on the basis of (*a*) hardenability, (*b*) close-limit composition, or (*c*) a combination of the two.

Reference has been made at various times to improved heat-treatment techniques and to the fact that the lean-alloy steels require much greater care in heat-treatment. Mr. Rohl mentioned that improvements in heat-treatment practice had occurred in the U.S.A. since the inception of boron steels. How have these improvements been made? According to the literature, the heat-treatment procedure in the U.S.A. seems to have varied very little in the last 10–15 years.

What happens to off-casts, when the hardenability of the steel is outside the limits set by the melter? Are they remelted, or are they sold to other customers who can utilize a steel having a different hardenability value?

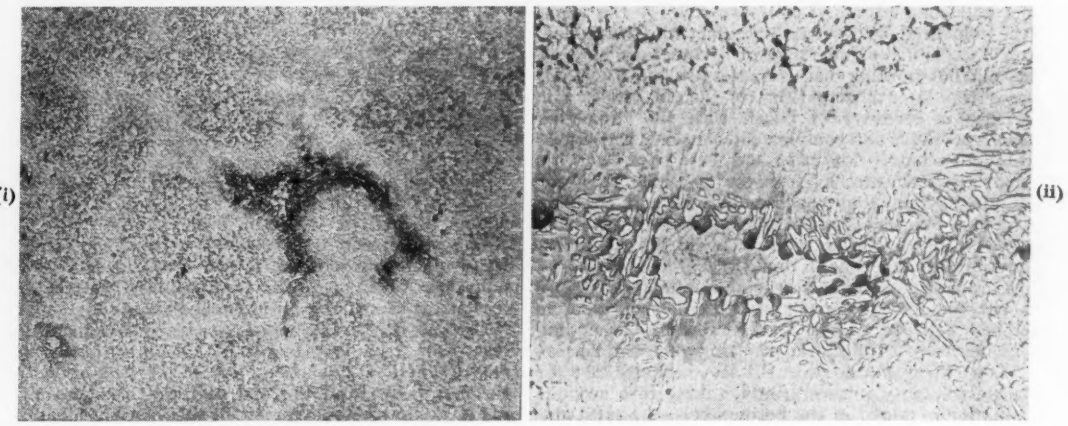


Fig. D—(i) Segregated area in the boron-treated low-alloy steel sand-casting, showing Fe–B–C eutectic; (ii) microstructure of Fe–B–C eutectic (i) × 100, (ii) × 500

In regard to the boron limit shown in the paper for additions of the complex alloys compared with additions of the simple alloys, I gathered that the yield from the complex alloys could, to be effective, be much lower than the yield from the simple alloys; yet Fig. 4 showed a similar boron content for steels made both with the complex alloys and with the simple alloys. Was the higher yield intentional, to overcome the undoubted danger of working to a minimum boron limit, when you might run into the band on the diagram where the effect of boron was unpredictable?

Does the addition of the excess of aluminium over the usual practice tend to produce ultra-fine grain size? Many of us during the war had experience of American steels with exceedingly fine grain sizes which proved difficult to machine. Is there any greater danger in these boron steels of these ultra-fine grain sizes occurring?

The saving of nickel in the U.S.A. in 1952 was said to be 12 million pounds. Is that the overall saving of nickel, or was it the weight of nickel saved by the adoption of boron steels as compared to the alloy steels without boron?

Are the boron steels with the lower alloy contents the same price as or cheaper than the steels they replace?

Mr. **Rohl:** In regard to the relationship between user and steelmaker in the selection and supply of steel, the chemistry limits as listed in the A.I.S.I. publications usually prevail, although the so-called 'H-bands' (hardenability bands) established for many grades of steel have become increasingly popular. These H-bands prescribe upper and lower limits of hardenability, under standard conditions of treatment, and provide slightly broader limits in chemistry ranges which, in turn, permit some degree of flexibility in balancing the chemical composition to meet final hardenability requirements.

In the matter of improvements in heat-treating practices, our reference relates largely to the broader understanding of the fundamentals of heat-treatment and the improvements that have occurred in modern heat-treating equipment.

The disposition of boron heats outside hardenability ranges has not been a problem, firstly, because the percentage of such heats is less than normally occur in furnishing steel to chemical composition ranges, and, secondly, because diversion outlets with suitable hardenability ranges have been found in most instances.

The actual percentage recovery of boron from the simple and complex alloys is approximately the same and is influenced by various manufacturing variables. Figure 3 depicts the full range of satisfactory hardenability and hot-working characteristics (0·0005–0·007%), whereas the later charts (Fig. 4) merely reflect our experience in recovery of boron with the respective additions of simple and complex alloys in the proportions used by our plants.

Recoveries from simple alloys, on the whole, have not been quite as consistent as those obtained with complex alloys, and for that reason the additions have been aimed at a slightly higher point to avoid the possibility of entering into the zone of doubtful hardenability.

The grain size is related to the aluminium added, and that will vary in different plants. There is considerable difference between plants and companies, depending on the degree of oxidation at the time of manufacture, etc. Our aim would be an A.S.T.M. grain size of 6 for most hardenability purposes; and, taking into account the aluminium added in the boron steels, the total aluminium addition is controllable, so that final grain-size results are comparable to steels of similar composition which do not contain boron.

The saving of 12,000,000 lb. of nickel in 1951 was reported as being related to the substitution of boron steels containing no nickel or lower percentages of nickel than the nickel-bearing grades formerly used.

The matter of price of boron steels as compared to those replaced is, of course, a commercial phase somewhat removed from the technical aspects. However, in considering the use of boron steels for non-bearing boron steels to given hardenability ranges, an overall saving is to be expected. Examples of savings, as reported by our Sales Department, reflect an advantage of **$12.00** per ton in the use of *TS*14*B*35 for 4140, **$25.00** per ton in the use of *TS*86*B*45 for 2340, and **$10.00** per ton in the use of *TS*43*BV*14 for 4815.

Mr. **T. W. Whiting** (Guest Keen and Nettlefolds (Midlands) Ltd.): We are not neglecting the conservation of alloys in the United Kingdom. During 1952 at the International Materials Conference in Washington the use of boron steel for bolts was advocated, and so the bolt industry in this country set out to use some boron steel, with the help of the Brymbo Steel Works.

All the usual alloys in the steel were eliminated except manganese and a cast of steel was made to the American *TS*10*B*37 specification. When this steel was subjected to the usual forms of heat-treatment (hardening at 850° C. and tempering at 550° C., with oil quenching in both cases), we obtained a tensile figure better by 10 tons than the normal figure for the same material (S.A.E. 1037) without boron; it would therefore be possible to use *TS*10*B*37 in place of S.A.E. 4037 and eliminate the use of molybdenum (*see* Table *F*). That is important, because large quantities of A.25 aircraft high-tensile bolts will be required in the 55–65-ton tensile range, and will be obtainable in a steel without using any of the usual alloys at all.

Table F

CHEMICAL COMPOSITIONS AND TENSILE STRENGTHS OF BORON AND NON-BORON BOLT STEELS

Steel	Approximate Chemical Composition, % C	Mn	Mo	B
***TS*10*B*30**	0·30	0·80	Nil	0·003
S.A.E.1037	0·37	0·80	Nil	Nil
S.A.E.4037	0·37	0·80	0·30	Nil
***TS*10*B*37**	0·37	0·80	Nil	0·003

B.S.S. Bolt Specification	Tensile Range, tons/sq. in.	Material Present	Future
1083/R	**45–55**	**S.A.E.1037**	***TS*10*B*30**
1083/R and A.25	**55–65**	**S.A.E.4037**	***TS*10*B*37**

The steel industry, both in Britain and in America, seems to be working towards the use of hardenability figures arrived at with the Jominy test piece. The Jominy test piece, with a diameter of 1¼ in., using the water-quenching method, is not much use to the bolt industry, which requires hardenability figures using test pieces less than ⅝ in. in diameter and oil-quenching.

Certain British Standards recognize a test using a test piece 3 in. long and ½ in. diameter, hardened at 850° C. and oil-quenched; this test piece is then cut through ½ in. from the end and the hardness test is made across there. For small sections this test is of more practical use than the Jominy test and will show the boron steel to advantage.

Boron should not be ignored for use in plain carbon

steels. Most bolts made to B.S.S. 1083/R now use the chemical composition range of S.A.E. 1037. It should be possible to lower the carbon content of the S.A.E. 1037 specification and bring it down to *TS10B30* specification, thereby providing softer wire in the spheroidized condition for cold-working afterwards (about 7 tons lower tensile strength and a much lower yield point). The softer wire would not crack so readily during subsequent cold-working and tool life would be greatly improved. The *TS10B30* steel would have the same final mechanical properties (45–55 tons/sq. in. tensile) as S.A.E. 1037 when subjected to the usual heat-treatment (hardening at 850° C. and tempering at 550° C., with oil-quenching in both cases).

Mr. **Rohl:** I would not advocate the use of boron if it is unnecessary, and if your sections are small you may be able to meet the properties without it. In the really low-alloy steels there are many instances where we have not seen any benefits from the use of boron, and so we have not used it. In the instance which has been given, however, there seem to be very good possibilities of gaining all the properties that are wanted.

REFERENCES

1. G. A. RUDOLPH and L. C. FLICKINGER: *Steel*, 1943, vol. 112, Apr. 5, pp. 114, 131, 149.
2. G. E. SPEIGHT: *J. Iron Steel Inst.*, 1950, vol. 166, pp. 196–198.
3. W. E. BARDGETT and L. REEVE: *Ibid.*, 1949, vol. 163, pp. 277–294.
4. R. A. GRANGE and T. M. GARVEY: *Trans. Amer. Soc. Metals*, 1946, vol. 37, pp. 136–174.

PROCEEDINGS OF THE AFTERNOON SESSION, 1.45 TO 4.45 P.M.

American Applications of Boron and Other Low-Alloy Steels

By H. B. Knowlton

SYNOPSIS

In the United States, to conserve nickel and molybdenum about 8–10% of the tonnage of low-alloy constructional steels has been converted to lower-alloy steels containing boron, with satisfactory results. Passenger-car factories using steels within government restrictions for alloying elements have not changed specifications. The International Harvester Company has used about 80,000 (short) tons of chromium–boron steels per year in the manufacture of trucks and tractors, some of which encounter very severe service. No failures attributable to the steel have been reported.

The paper explains why this has been possible. American practice is based on successful performance of finished parts. Breakage may be due to either lack of strength or lack of toughness. Strength at each point on the cross-section must be higher than the stress encountered. Except for a small percentage of parts subject to pure tension, the stress is maximum at, or near, the surface and zero at the neutral axis. High strength at the neutral axis is useless. Strength at different points on the cross-section is dependent upon hardenability. During the last war, U.S. industry created a series of low-alloy steels on the basis of duplicating the hardenability of ones previously used. These gave satisfactory performance in most cases. Boron steels have been proposed for heat-treated parts on the same basis.

Boron steels have shown adequate toughness for safe performance of heat-treated parts for which they have been tried. This includes light and heavy coil and leaf springs, and axles up to at least 3 in. dia. Notched-bar impact strength in the centre of large sections does not seem to correlate with service performance. Tests of finished parts have shown as good fatigue strength for boron steels as for other alloy steels of equivalent hardenability.

Boron steels have not been so satisfactory for case hardening, owing to the difference in the effect of hardenability on the case and the core. Some trouble with distortion has been reported. Nevertheless, a fair percentage of case-hardened gears has been satisfactorily made from certain boron steels. 779 B

Introduction

THE alloy steels containing nickel, chromium, and vanadium have become increasingly popular for two general types of application: (i) for oxidation and corrosion resistance, where the chemical properties of the alloying elements are of maximum importance, and (ii) in a large variety of automotive, aircraft, and other applications, where high mechanical properties such as strength, toughness in proportion to weight, etc., are desired. In type (i), where heat and corrosion resistance are essential, it may be difficult to find an abundant supply of other elements that would duplicate the chemical properties of alloys of chromium, nickel, and molybdenum, although something has been accomplished in decreasing the severity of temperature and other operating conditions. Great progress has been made, however, in the conservation of these alloying elements in the manufacture of steel parts requiring high mechanical properties. For example, the International Harvester Company has been successful in replacing about 70% of its annual tonnage of alloy steels containing nickel and molybdenum by a series of chromium–boron steels, without impairing in any way the performance of its products, which include trucks and tractors of all sizes, some of them operating under extremely severe conditions—*e.g.*, large tractors replace dynamite in knocking rocks out of mountain sides.

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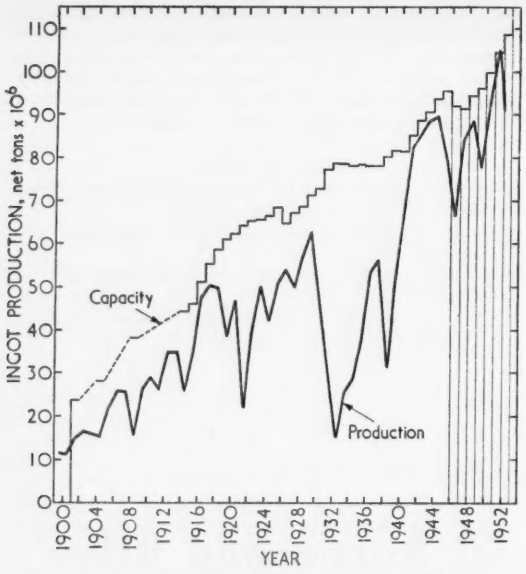

Fig. 1—Rise in U.S. ingot production, 1899–1953

About two years ago, the Society of Automotive Engineers set up Division VIII under their Iron and Steel Technical Committee, to study boron and other low-alloy steels. At first it was intended to have a committee of only 10 or 12 members, but the interest in the subject was so great and the demands for permission to attend meetings, at least as spectators, were so numerous that open forum discussions are now held every two or three months with attendances of 100–150.

Whilst there have been, and still are, some differences of opinion, considerable progress has been made in the conservation of alloying elements. It is true that only about 9% of the alloy steel production has been converted to boron types of alloy. It must be remembered, however, that in the passenger-car industry, where section sizes are relatively light, it has been common practice to use low-alloy steels such as 4000, 1300, and 5100, all of which come within the present Government restrictions so far as nickel and molybdenum are concerned; consequently, further conservation is not necessary. The American Society for Metals has had several committees working on this problem, and both societies have published data on the boron steels. Figures 1 and 2 show the rise in American ingot production in the last 50 years, and Fig. 3 shows the boron-steel requirements of this Company and the U.S.A. in 1952.

The Earlier Low-Alloy Steels

Before explaining the American plan for the selection and application of boron steels as alternatives for the previous standard alloy constructional steels, it may be as well to discuss briefly the evolution of the old types of steel. These steels were developed very largely by the automotive industry, and are commonly known as the S.A.E. (Society of Automotive Engineer) steels. The automotive practice of selection

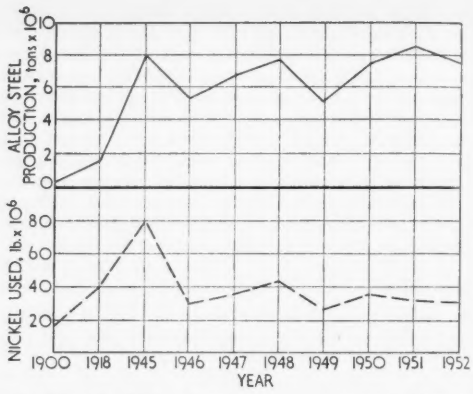

Fig. 2—Rise in U.S. alloy steel production, 1900–1952

of steel and writing of specifications for mechanical or physical properties of finished parts grew up, to some extent at least, by trial and error. The necessity for maximum strength in proportion to weight made it desirable to consider a reduction of sizes in all new designs. On the other hand, any difficulties with field performance usually led the engineer to increase section size, or to demand higher hardness and strength of the critical part. Eventually, a combination of composition and mechanical properties was attained which fairly well guaranteed successful performance with minimum weight, and sometimes with minimum cost.

The selection of alloy content, however, was far from uniform throughout the industry. There was a tendency to attribute specific properties, such as high toughness in proportion to strength, to certain alloy steels. Various automotive manufacturers standardized on different types of alloy, and even now some metallurgists feel that a particular type is necessary for the safe performance of their product. Nickel, chromium, and molybdenum, when alloyed with steel, produce varying degrees of heat-, oxidation-, and corrosion-resistance which depend upon the chemical composition of the steel. On the other hand, it has been found that physical properties, such as hardness, strength, and toughness, can be produced with various types of alloy steel; also, in the great majority of cases it is possible to change the composition from one type of alloy to another and at the same time duplicate all of the physical properties of the original steel. In 1942 it was found that most of the new types of alloy steel that were set up on the basis of

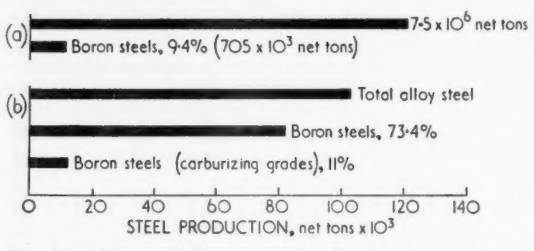
(a)
7·5 x 10⁶ net tons
Boron steels, 9·4% (705 x 10³ net tons)
(b)
Total alloy steel
Boron steels, 73·4%
Boron steels (carburizing grades), 11%
STEEL PRODUCTION, net tons x 10³

Fig. 3—Alloy and boron usage in the International Harvester Company and the U.S.A. in 1952

duplicating the hardenability of the previous steels gave satisfactory performance. There were, however, some exceptions. Boegehold and others have very strongly urged the study of all critical parts, on the basis of simulated service performance tests applied to the part itself.

It must always be remembered that the first criterion is the satisfactory performance of the individual part. The ideal goal in conservation studies is to determine the minimum amount of critical alloys that is required to guarantee satisfactory performance.

Another outstanding factor in the development of American specifications has been the large continuous heat-treating furnaces. Economical operation of these furnaces depends upon a continuous flow of work through them, with the minimum changes in quenching and tempering temperatures. This places an extra premium on control of hardenability.

Table I shows the common types of S.A.E. steels and the four-digit numbering system for designating them. The first two digits indicate the general type, such as Ni, Ni–Cr, Cr–Mo, etc. The last two figures designate the mean carbon content of the particular steel, and this is usually specified within a range of 0·5% for low-carbon steels. Before 1941 all these steels were considered necessary, but the 3·50% and 5·00% Ni steels and the 3100 Ni–Cr steels have now been almost completely eliminated. The 8600, 8700, and 9000 series of low triple-alloy steels, which were created as National Emergency Steels after the start of the 2nd World War, have become some of the most popular of the automotive alloy steels. This applies particularly to the 8600 steels and the low-carbon 9400 case-hardening types. More recently, Government restrictions on nickel and molybdenum have made it advisable to set up certain 'Temporary Standard' or 'TS' modifications of some of these steels.

Table I

S.A.E. NUMBERING SYSTEM FOR WROUGHT OR ROLLED STEEL

Type of Steel	Numerals and Digit
Carbon Steels	1xxx
Plain carbon	10xx
Free-cutting (screw stock)	11xx
Manganese Steels	
1·75% Mn	13xx
Nickel Steels (obsolete)	2xxx
3·50% Ni „	23xx
5·00% Ni „	25xx
Nickel–Chromium Steels	3xxx
1·25% Ni, 0·65 or 0·80% Cr (obsolete)	31xx
3·50% Ni, 1·55% Cr	33xx
Molybdenum Steels	
0·25% Mo	40xx
Chromium–Molybdenum Steels	
0·50 and 0·95% Cr, 0·25 and 0·20% Mo	41xx
Nickel–Chromium–Molybdenum Steels	
1·80% Ni, 0·50 and 0·80% Cr, 0·25% Mo	43xx
0·55% Ni, 0·50% Cr, 0·20% Mo	86xx
0·55% Ni, 0·50% Cr, 0·25% Mo	87xx
3·25% Ni, 1·20% Cr, 0·12% Mo	93xx
0·45% Ni, 0·40% Cr, 0·12% Mo	94xx
0·55% Ni, 0·17% Cr, 0·20% Mo	97xx
1·00% Ni, 0·80% Cr, 0·25% Mo	98xx
Nickel–Molybdenum Steels	
1·55 and 1·80% Ni, 0·20 and 0·25% Mo	46xx
3·50% Ni, 0·25% Mo	48xx
Chromium Steels	5xxx
Low-chromium: 0·27 and 0·65% Cr	50xx
High-chromium: 0·80, 0·95, or 1·05% Cr	51xx
Chromium–Vanadium Steels	
0·95% Cr, 0·15% V (min.)	61xx
Silicon–Manganese Steels	9xxx
0·65 and 0·85% Mn, 1·40 and 2·00% Si	92xx

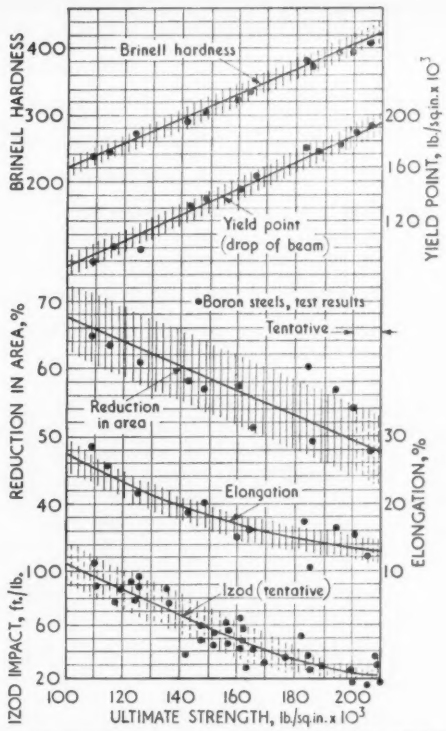

Fig. 4—Correlation between tensile strength and other physical properties of fully heat-treated low-alloy steels containing 0·30–0·50% C (Wray)

Performance

It cannot be emphasized too strongly that successful performance is the fundamental criterion. If the maximum conservation of alloying elements is to be effected without impairing the performance of the product, the question of how and why any particular part succeeds or fails must first be determined. Among the common types of failure and their remedies may be mentioned the following:

Type of Failure	Remedy
Corrosion	Change chemical composition, or apply corrosion-resistant coating
Wear	Usually requires higher hardness

Type of Failure	Remedy
Temporary deflections	Redesign necessary, as composition and heat-treatment do not change modulus of elasticity
Permanent deflection	Raise elastic limit or yield strength. This usually means an increase of hardness
Breakage	Discussed below.

The three major causes of breakage are:

(i) Lack of strength, at some points on the cross-section, to withstand the stress that appears at that location

(ii) Lack of sufficient toughness to withstand plastic deformations that must take place

(iii) Undesirable internal stresses, which decrease the working strength of the part.

Figure 4 shows the correlation between tensile strength and other physical properties of various types of automotive alloy steels, including the boron steels. The Brinell hardness and yield strength remain very closely proportional to the tensile strength. All the points fall within very narrow bands. Similarly, the elongation, reduction in area, and Izod notch toughness are approximately inversely proportional to tensile strength, and again, the points representing observed properties of boron steels fall within the bands of normal expectancy.

PHYSICAL PROPERTIES OF BORON AND LOW-ALLOY STEELS

The compositions of the steels referred to in this paper are given in Table II. The A.I.S.I. steels are designated by four digits, the first two indicating the type of steel (C, Ni, Ni–Cr, etc.) and the last two giving the mean carbon content in $\frac{1}{100}$%. The letter *B* between two groups of digits indicates that the steel contains boron, and the letter *H* indicates a steel ordered to a hardenability specification instead of to a chemical specification.

Impact Strength

The most frequently occurring question relating to boron steels has been in regard to toughness or impact strength. This question could probably be answered most simply by stating that, in general, the boron steels, when properly heat-treated, will have the same degree of impact strength as the steel being replaced. However, further discussion on this point is necessary. It is known that any part of the steel requires a certain amount of toughness or ability to withstand plastic deformation, but the degree of toughness required is often much lower than is realized. There has been a tendency to determine the toughness of the steel in the original part and to establish this as a criterion for alternative steels for such a part, regardless of whether the toughness is actually essential to safe performance.

The most common method of measuring toughness is the impact test, which applies a hammer blow to a small specimen containing a **V** notch, the results being expressed in terms of foot-pounds of energy absorbed in breaking the specimen. This is frequently called the impact strength, although it would be more accurate to describe it as the notch toughness.

Table II

CHEMICAL COMPOSITION OF A.I.S.I. AND S.A.E. STEELS

A.I.S.I. or S.A.E. Number	C, %	Mn, %	Ni, %	Cr, %	Mo, %	B (min.), %	Other Elements, %
13xx	Variable	1·60–1·90	...	...	...	...	
31xx	Variable	0·40–0·60 0·60–0·80 0·70–0·90	1·10–1·40	0·55–0·75	...	...	
3312	0·09–0·14	0·45–0·60	3·25–3·75	1·40–1·75	...	...	
40xx	Variable	0·70–0·90	...	...	0·20–0·30	...	
4140	0·38–0·43	0·75–1·00	...	0·80–1·10	0·15–0·25	...	
TS43BV12	0·08–0·13	0·75–1·00	1·65–2·00	0·40–0·60	0·20–0·30	0·0005	Va, 0·03 (min.)
TS43BV14	0·10–0·15	0·45–0·65	1·65–2·00	0·40–0·60	0·08–0·15	0·0005	Va, 0·03 (min.)
4620	0·17–0·22	0·45–0·65	1·65–2·00	...	0·20–0·30	...	
4815	0·13–0·18	0·40–0·60	3·25–3·75	...	0·20–0·30	...	
50B20	0·17–0·23	0·70–1·00	...	0·35–0·60	...	0·0005	
50B50-H	0·47–0·55	0·70–1·00	...	0·35–0·60	...	0·0005	
50B60-H	0·55–0·65	0·70–1·00	...	0·35–0·60	...	0·0005	
51B20	0·17–0·22	0·70–0·90	...	0·70–0·90	...	0·0005	
5152	0·48–0·55	0·70–0·90	...	0·90–1·20	...	...	
51B60	0·55–0·65	0·70–1·00	...	0·70–0·90	...	0·0005	
7045	0·42–0·49	0·90–1·20	...	...	0·30–0·40	...	
80B20	0·17–0·23	0·60–0·90	0·20–0·40	0·15–0·35	0·08–0·15	0·0005	
8442	0·38–0·42	1·30–1·60	...	...	0·30–0·40	...	
8620	0·18–0·23	0·70–0·90	0·40–0·70	0·40–0·60	0·15–0·25	...	
8640	0·38–0·43	0·75–1·00	0·40–0·70	0·40–0·60	0·15–0·25	...	
8650	0·48–0·53	0·75–1·00	0·40–0·70	0·40–0·60	0·15–0·25	...	
8650-H	0·46–0·54	0·70–1·05	0·35–0·75	0·35–0·65	0·15–0·25	...	
87xx	Variable	0·60–0·80 0·75–1·00	0·40–0·70	0·40–0·60	0·20–0·30	...	
9260	0·55–0·65	0·70–1·00	...	...	...	...	Si, 1·80–2·20
9262-H	0·55–0·65	0·70–1·05	...	0·20–0·50	...	...	Si, 1·70–2·20
94xx	Variable	0·90–1·20 1·00–1·30	0·30–0·60	0·30–0·50	0·08–0·15	...	
94B17	0·15–0·20	0·75–1·00	0·30–0·60	0·30–0·50	0·08–0·15	0·0005	

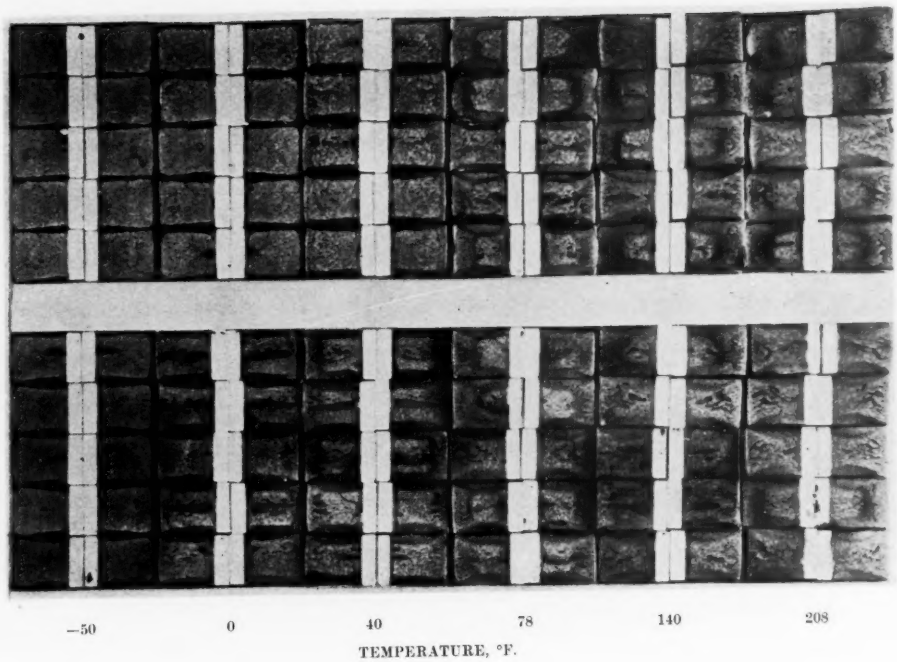

Fig. 5—Fractures of impact specimens of a straight Cr steel (5152) and an experimental ingot of low Cr–Mo steel (*X*-1066)

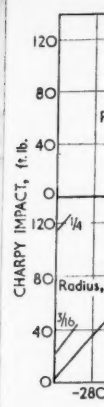

Fig. 6—Ef
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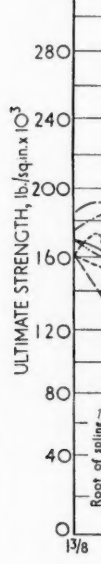


Fig. 7—Te

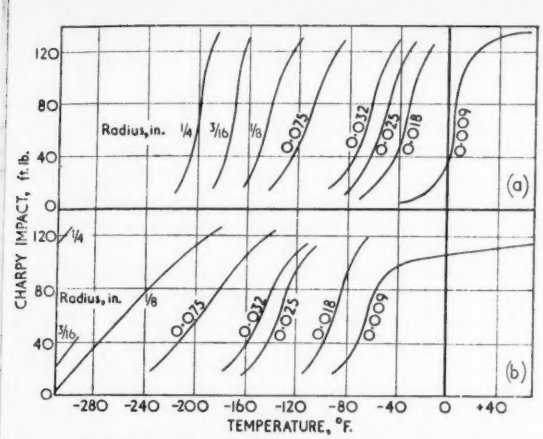


Fig. 6—Effect of notch sharpness on the impact properties of (*a*) normalized 0·15% C steel, (*b*) normalized 0·15% C, 2·00% Ni steel (Wray)

To make the impact, or notch toughness, test more drastic it is common practice to make tests at a series of temperatures. Any steel is inclined to become more brittle with decrease of temperature, increase of sharpness of notch, and increase of velocity of application of stress, particularly where high velocities such as those produced in ballistics are involved. The transition temperature from tough to brittle performance varies with the type of steel involved. Figure 6 shows the effect of the size of the fillet at the bottom of the notch upon the observed impact strength of two steels. The transition temperature for the plain carbon steel (Fig. 6*a*) is about 0° F. (— 18° C.), but it is lowered to less than — 120° F.

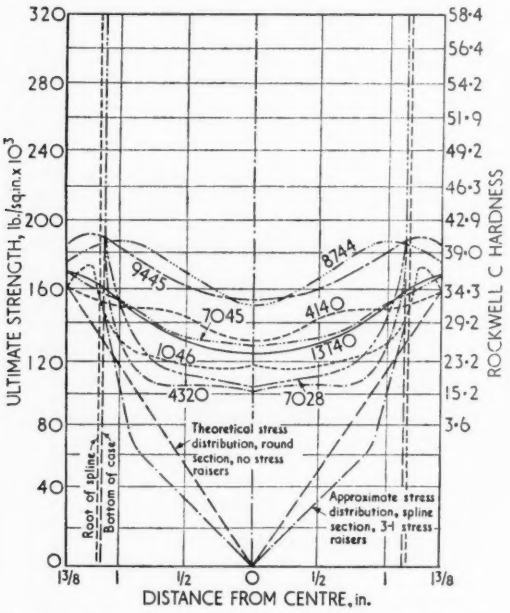


Fig. 7—Tensile strength of 2¾-in. dia. axle shafts of different steels

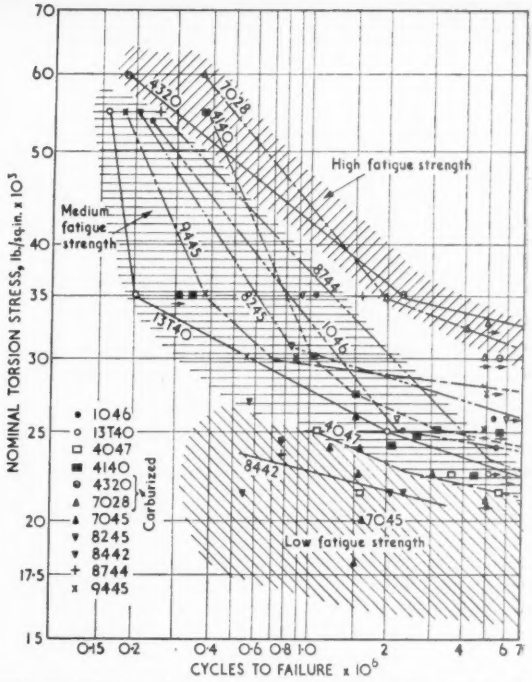


Fig. 8—Torsion fatigue tests of axle shafts in Fig. 7

(— 84° C.) by increasing the radius of the fillet to a more commercial value of ⅛ in.

Figure 5 shows the fractures of impact specimens of two types of spring steel (5152, a straight Cr steel, and *X*-1066, an experimental low Cr–Mo steel) tested at various temperatures; the fractures show a change from tough to brittle appearance when the testing temperature is lowered below the transition temperature. The transition appears at a very much lower temperature with the experimental ingot. On the basis of these tests alone, there might be a tendency to decide that the 5152 steel could not be used for automotive leaf springs where operating temperatures are below 0° F. In fact, however, this type of steel has been used for this purpose for several years by two of the largest automotive manufacturers in the world. The final spring does not contain a sharp notch, and consequently does not require a high degree of cold notch toughness, despite the fact that the spring may encounter a great deal of very rough service at sub-zero temperatures.

Hardness

As mentioned earlier, the hardness and strength at each point over the cross-section of a finished part must be higher than the stress that will be encountered at that point. In a large axle shaft, subject to either bending or torsion, the maximum stress appears at the surface, while there is a locality of zero stress at the neutral axis, which is usually near the centre of the shaft.

Figure 7 shows the cross-sectional hardness and tensile strength of 2¾-in. dia. tractor axle shafts made from different types of steel. The maximum stress

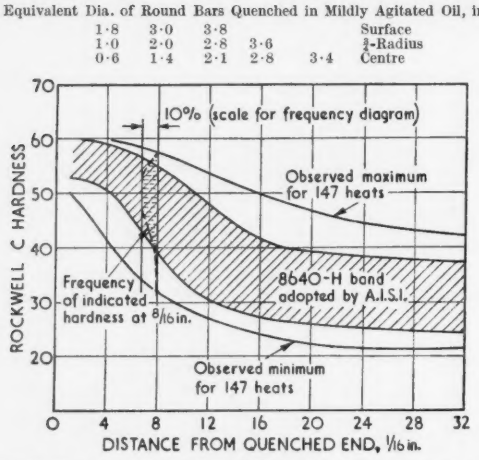

Equivalent Dia. of Round Bars Quenched in Mildly Agitated Oil, in.
1·8 3·0 3·8 Surface
1·0 2·0 2·8 3·6 ¾-Radius
0·6 1·4 2·1 2·8 3·4 Centre
10% (scale for frequency diagram)
Observed maximum for 147 heats
8640-H band adopted by A.I.S.I.
Frequency of indicated hardness at 8/16 in.
Observed minimum for 147 heats
ROCKWELL C HARDNESS
DISTANCE FROM QUENCHED END, 1/16 in.

Fig. 9—Results of hardenability tests on 147 heats of 8640 steel

that can appear at the surface without causing a failure at that point is less than 160,000 lb./sq. in. This corresponds with the minimum specified Brinell hardness (321 B.H.N.). Assuming a neutral axis, or locality of zero stress, at the geometric centre, the distribution of stress can be represented by a straight line from 160,000 lb./sq. in. at the surface to zero at the centre. To ensure satisfactory performance it is necessary only that the cross-sectional hardness and strength curves should fall above the expected stress line. In practice, stress raisers such as splines and keyways may produce concentrations of stress towards the surface, so that the stress-distribution line is a curve rather than a straight line (*see* Fig. 7).

Fatigue Tests

The results of torsion fatigue tests of these axle shafts are given in Fig. 8. Although considerable experimental error is involved in this type of test, all the results obtained fall within a conventional *SN* band with the exception of the 8442 and 7045 steel axles, which tend to give poor performance, particularly at the higher stresses. Not much use was made of these two Mn–Mo steels (which contained over 1% of Mn), but no field failures were experienced with the axles machined from them.

The band at the top of Fig. 8 indicates the results of torsion fatigue tests of two carburizing steels, after conventional case-hardening treatments of the axles. The fatigue performance is considerably better where there is a high surface hardness and strength and a relatively low centre hardness. It is clear, therefore, that a necessary property of the selected steel is that the strength at different points on the cross-section should be higher than the stress to be encountered; but it is not necessary to demand higher degrees of strength and hardness than are needed to meet the above requirement. Specifications for higher strength and hardness at the centre, or half-radius, than are needed to withstand the stresses at these points have been the principal causes of the wasteful use of alloy steels, and are the principal obstacles to overcome in substituting boron steels for the conventional alloys.

Hardenability

It is impossible to discuss the American use of boron steels without first considering the American practice of hardenability specifications. The term 'hardenability' in the U.S.A. means the 'ease' of hardening, not the actual degree of hardness. Any steel containing more than 0·40% of carbon can be made very hard (over 60 R_C) by drastic quench. However, the speed of quench required to attain this hardness varies with different types of steel. In general, the higher the alloy content the lower will be the quenching speed that is necessary to attain full hardness. The Jominy, or end-quench hardenability, test has been found to be the most convenient method for determining the hardenability of different types of steel.

With any given type of steel it is necessary to permit a range of composition for each individual element. It would obviously be impossible for all batches of steel made within such ranges to have identical hardenability. By plotting hardenability tests of samples taken from 100 or more heats of steel, it is possible to establish a hardenability band that fairly accurately represents the entire spread of hardenability which may be expected with that type of steel. This is illustrated in Fig. 9 (for 8640 steel), in which each vertical line on the chart represents a different speed of cooling. These can be translated approximately into speeds of cooling at the surface or at some point on the cross-section of round bars of different diameter, after quenching in oil or water.

For example, in Fig. 9 the vertical line at $\frac{4}{16}$ in. from the quenched end represents the approximate cooling speed of the surface of a 1·8-in. dia. bar, or the $\frac{3}{4}$-radius position of a 1·6-in. dia. round bar, after conventional oil-quenching. Thus, the maximum and minimum as-quenched hardness that can be expected with oil- or water-quenching of different size rounds

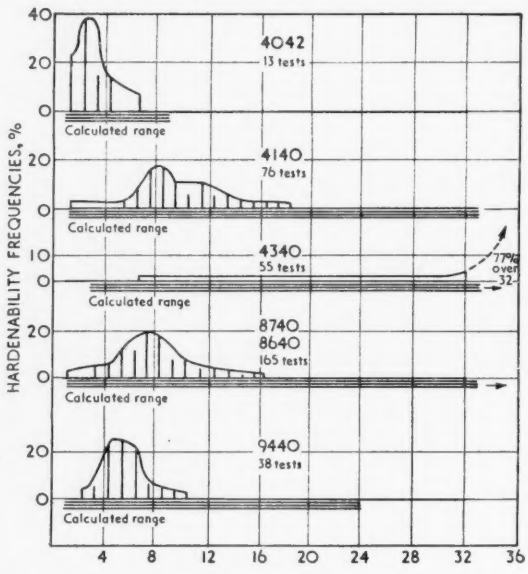


Fig. 10—Hardenability frequencies of steels at 0·50% C level

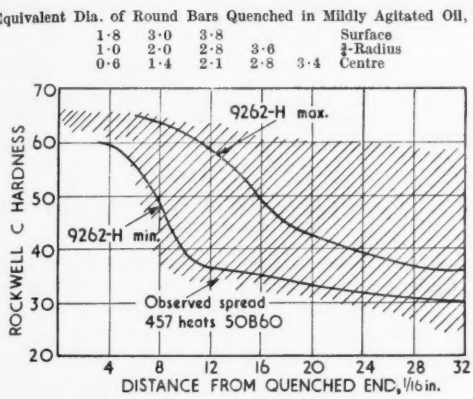

Equivalent Dia. of Round Bars Quenched in Mildly Agitated Oil, in.
1·8 3·0 3·8 Surface
1·0 2·0 2·8 3·6 ¾-Radius
0·6 1·4 2·1 2·8 3·4 Centre
ROCKWELL C HARDNESS
9262-H max.
9262-H min.
Observed spread
457 heats 50B60
DISTANCE FROM QUENCHED END, 1/16 in.

Fig. 11—Results of hardenability tests on 457 heats of 50*B*60 Cr–B steels

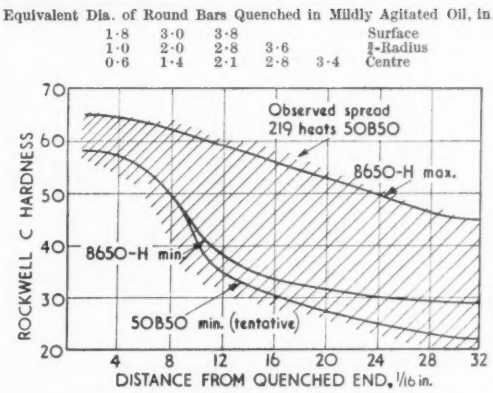

Equivalent Dia. of Round Bars Quenched in Mildly Agitated Oil, in.
1·8 3·0 3·8 Surface
1·0 2·0 2·8 3·6 ¾-Radius
0·6 1·4 2·1 2·8 3·4 Centre
ROCKWELL C HARDNESS
Observed spread
219 heats 50B50
8650-H max.
8650-H min.
50B50 min. (tentative)
DISTANCE FROM QUENCHED END, 1/16 in.

Fig. 13—Results of hardenability tests on 219 heats of 50*B*50-*H* steel

of any given type of steel can be predicted from the hardenability band. The most important concern is probably the minimum hardness, as this also indicates the minimum strength at different points on the cross-section, which must be greater than the stress to be encountered. The steel producers have said they would be willing to guarantee the hardenability of any given type of steel within a slightly narrower band than the total observed spread. The narrow specified band is shown by the shaded area in Fig. 9.

This curve also shows the frequency at which different degrees of hardenability appeared within the 147 heats plotted. The width of the band can be materially reduced by eliminating approximately 3% of the heats having the lowest hardenability, and another 3% of the heats having the highest. The present hardenability, or *H* band, specifications were created on this basis.

Figure 10 shows the hardenability frequencies of several steels, plotted horizontally. It was formerly believed that each type of steel had radically different properties, but the frequency curves indicate that, with the exception of 4340, all the steels have approximately the same hardenability. The most important point, however, is the minimum hardenability, as this governs the minimum hardness that will be produced at various points on the cross-section of a given part

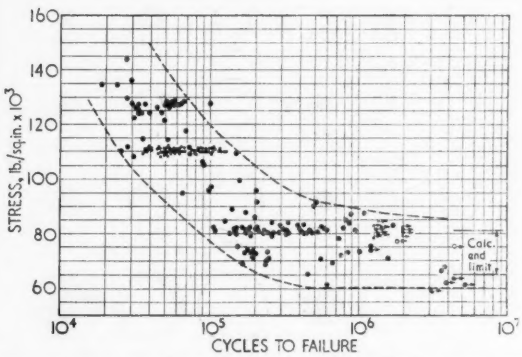


Fig. 12—Results of single leaf-bending fatigue tests on truck leaf springs (1·750 in × 0·323 in.)

by a conventional heat-treatment. It will probably be conceded that any part failing in service 2–3% of the time, or even failing to attain the minimum specified hardness for a similar period, will call for a change of the material specification, or change of heat-treatment. Therefore, the elimination of the 3% of the heats having the lowest hardenability overcomes most troubles arising in heat-treatment and in service.

The shaded area in Fig. 11 shows the observed spread of 457 heats of 50*B*60 chromium–boron steels, which are being used in place of 9260 for leaf springs. The solid lines represent the maximum and minimum specified hardenability for 9262-*H*. The minimum hardenability is as good as that of the 9260 and 9262-*H* steels previously used for springs of different thickness. The maximum hardenability is greater, but has not caused any difficulty. The 50*B*60 steel has been used for truck springs and tractor springs, up to about $\frac{3}{4}$ in. in thickness, with great success.

A typical fatigue, or *SN*, band for truck springs is given in Fig. 12. The results for the boron steels fall within the experience band for other types of steel. For heavy tractor springs over 1 in. thick, a similar steel (51*B*60) with a slightly higher chromium content has been used. No difficulties have been reported with the boron steels used in these applications.

Figure 13 shows the observed hardenability band for 50*B*50-*H* steel (shaded area), in comparison with steel 8650-*H* previously used for truck and tractor axle shafts. The minimum hardenability of the 50*B*50 steel is as good as that of the 8650 steel up to about $\frac{6}{16}$ in. from the quenched end of the Jominy specimen. This corresponds with the centre of a 1-in. oil-quenched round, the $\frac{3}{4}$-radius position of a 1·6-in. round, or the surface of a 2·5-in. round, which in turn corresponds with the critical location on the cross-section of different types of parts. Bolts are usually considered to be stressed uniformly over the entire cross-section, and therefore the hardness at the centre is taken as the criterion. This would mean that the 50*B*50 steel with conventional oil-quenching heat-treatment should give good results for bolts up to 1-in. dia. With axle shafts having roots of splines or keyways about one-quarter of the way from the surface to the centre (or at the $\frac{3}{4}$-radius position) this

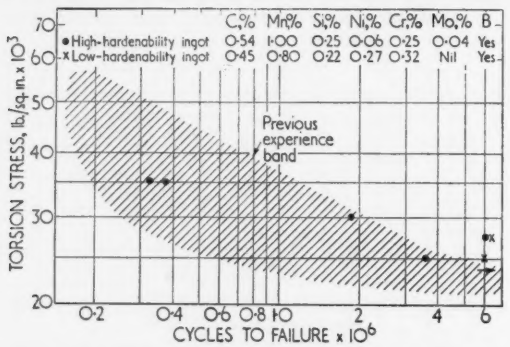

C,% Mn% Si,% Ni,% Cr,% Mo,% B
• High-hardenability ingot 0·54 1·00 0·25 0·06 0·25 0·04 Yes
x Low-hardenability ingot 0·45 0·80 0·22 0·27 0·32 Nil Yes
Previous experience band
TORSION STRESS, lb./sq.in. x 10³
CYCLES TO FAILURE x 10⁶

Fig. 14—Torsion fatigue tests on 2¾-in. dia. tractor axles of 50*B*50-steel

steel should be satisfactory for oil-quenched shafts up to 1·6-in. dia. and water-quenched shafts up to 2·9-in. dia. This corresponds very closely with our practice. With pressure-quenching, even larger sizes could be used.

The minimum section of a shaft may sometimes be cylindrical, in which case the surface hardness becomes of maximum importance; in this instance the 50*B*50 steel would be expected to give entirely satisfactory results up to 2·5-in. dia.

It will be noted from Fig. 13 that the majority of heats of 50*B*50 will be entirely satisfactory even for larger sizes. It is usual practice for the steel mill to make a hardenability test of each heat of steel before it is shipped to the customer. The small percentage of heats falling on the bottom of the hardenability spread are not applied to parts larger than those indicated. Although it is not yet possible to obtain many of the boron steels on hardenability bands, the intelligent application of heats has eliminated difficulties even in the larger sizes. Some users, however, prefer to limit the application of boron steels to sections not over 3 in. dia.

Fatigue Strength

Figure 14 shows the results of torsion fatigue tests on 2 3/16-in. dia. 50*B*50 tractor axles, in comparison with the previous experience bands on 8640 and other similar alloy steels. All the points plotted for the boron axles fall within or above the previous experience band for fatigue strength.

While on the subject of fatigue strength, it is worth noting that failure to produce 100% (or even 90%) martensite by quenching is not as serious, so far as fatigue strength is concerned, as it is in regard to impact strength, or notch sensitivity. Figure 15 shows the results obtained by Boegehold on specimens of 4140 steels, quenched for varying degrees of hardness and all tempered to 35 R_C. The specimen quenched for only 40 R_C showed an endurance limit of 82,000 lb./sq.in., compared

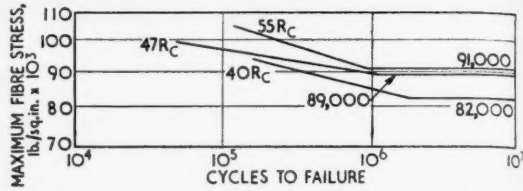


Fig. 15—Moore fatigue test on 2¾-in. dia. bars of 4140 steel (Boegehold)

with 91,000 lb./sq. in. for the specimen fully quenched for 55 R_C—a loss of only about 10% for slack quenching. It has frequently been stated that, for the boron steels, it is essential that the steel be quenched for a high percentage of martensite, so as to develop the normal impact properties. This may be true where impact strength or notch sensitivity is a predominant requirement, but it does not seem to be so important where only fatigue is involved.

Figure 16 shows the maximum diameters of oil-quenched rounds, of different types of steel, which can be depended upon to produce varying percentages of martensite at the surface, ¾-radius, and centre. If 90% martensite at the centre is the criterion, the minimum hardenability of 8640 steel would only guarantee 90% martensite to 0·7 in. maximum dia. Actually, this steel is used in sections up to 3 in. dia., which would imply that many parts are in service with only 50% martensite at the surface. Although it may be ideal to demand 90% martensite, at least at ¾-radius, there is good reason to believe that this is not always essential. Boegehold's work would indicate a decrease of about 10% in fatigue strength of unnotched pieces. Dolan has shown a decrease of about 20–25% in notch fatigue strength. If the design stress is held within these limitations, the poor type of structure may not cause any difficulty in service.

APPLICATION OF BORON STEELS

The effect of size upon the properties of 8640 steel is shown in Fig. 17. The hardness and strength at the ¾-radius position decrease for sizes over 1-in. in

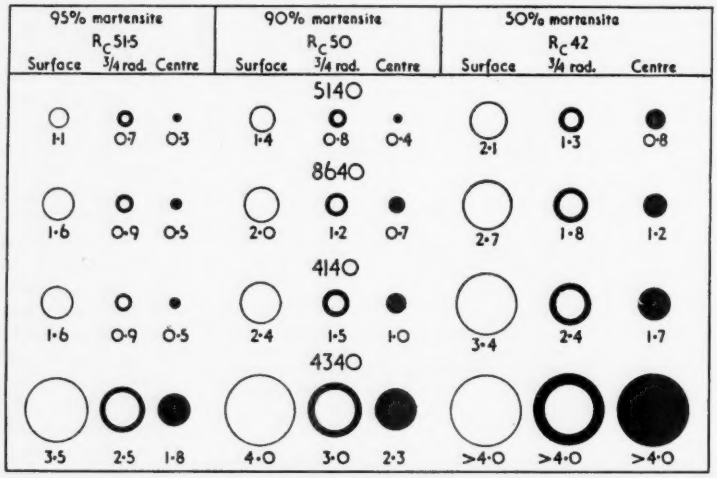


Fig. 16—Maximum diameters for oil-quenched rounds

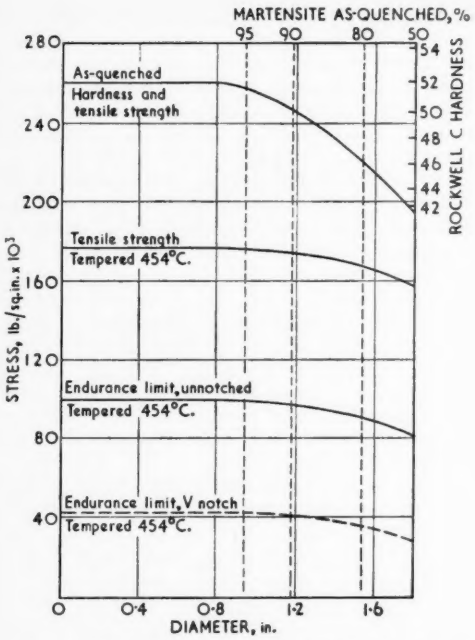


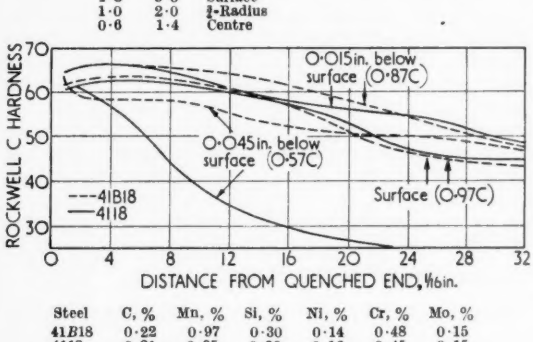
Fig. 17—Estimated minimum properties at ¾-radius of oil-quenched bars of 8640 steel

dia. After tempering at 454° C., however, the reduction of tensile strength in the larger sizes is less marked. The endurance limit in fatigue for the notch specimens drops from about 40,000 to about 30,000 lb./sq. in. This is based on nominal stress, not including the effect of stress raisers, such as fillets.

Case-Hardened Gears

The boron steels have not been so satisfactory for case-hardened gears, although several types of gears have been successfully made from 43*BV*12 and *BV*14, 94*B*17, 80*B*20 and 50*B*20 steels.

Boron has a much greater effect upon the hardenability of low-carbon steel than high-carbon steel. This is illustrated in Fig. 18, which shows the hardenability at different levels in a steel with and without boron,



Steel	C, %	Mn, %	Si, %	Ni, %	Cr, %	Mo, %
41*B*18	0·22	0·97	0·30	0·14	0·48	0·15
4118	0·21	0·95	0·30	0·16	0·45	0·15

Fig. 18—Hardenability of 4118 and 41*B*18 carburizing steels (Sailer)

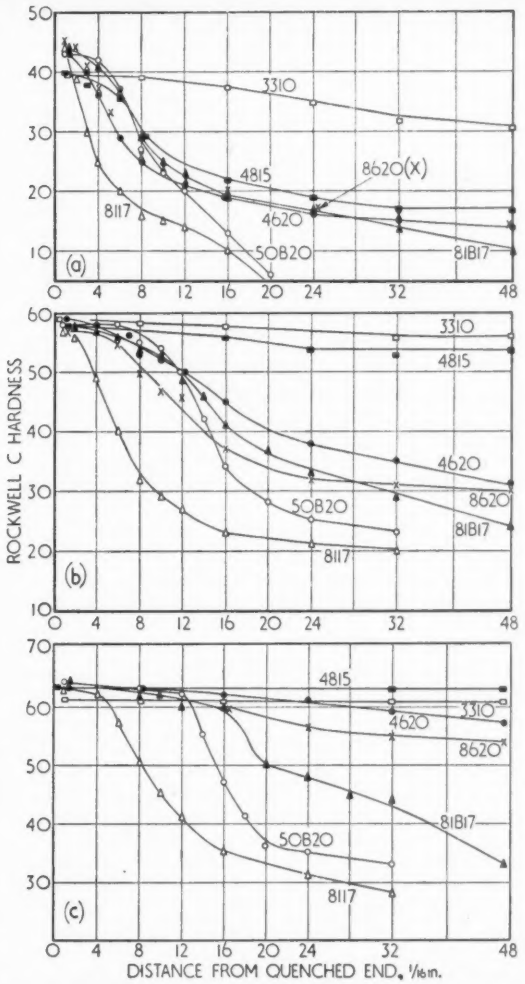


Fig. 19—Hardenability of seven commercial carburizing steels end-quenched from 1700° F. (925° C.): (*a*) Core hardenability; (*b*) case hardenability at 0·40% C level; (*c*) case hardenability at 0·60% C level (Thomson, Boegehold, and Kayser)

and even more clearly in Fig. 19. The hardenability of the cores of 8620, 4620, 4815, and 3312 steels are all about the same, but in the portion of the case that contains about 0·40% of carbon (Fig. 19*b*) the hardenability of 3312 and 4815 is much higher than that of the other steels. At 0·60% C (Fig. 19*c*), the 4620 steel is beginning to catch up with the 4815. Although the steels just mentioned do not contain boron, these data do show how the hardenability of different levels of the case may not be the same for steels having the same core hardenability.

In general, it may be stated that the substitution of boron for other alloys in a carburizing steel may easily maintain, or even increase, the hardenability of the core above that of the original steel, but that the hardenability of the outer layer of the case will be that of the *basic* composition of the new steel without boron; in other words, it may be lower than

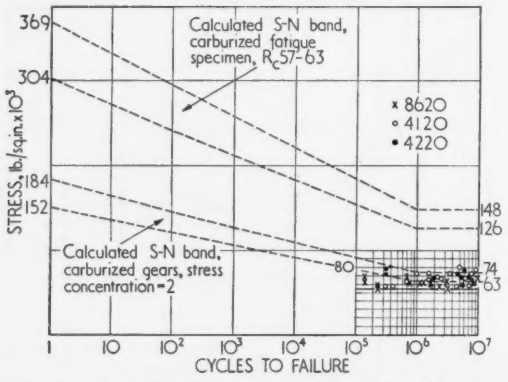


Fig. 20—Bending fatigue tests on carburized gear

that of the original higher-alloy steel. In small sections this may not be important, but in larger sizes it may lead to insufficient surface hardness of case-hardened parts.

On the other hand, increased hardenability in the core of the boron case-hardening grades has sometimes been a disadvantage, and may produce increased distortion. Some metallurgists attribute this to the difference between the hardenability of case and core. The 43*BV*12 and *BV*14 grades have proved very satisfactory substitutes for 4815 and 3312, for certain heavy-duty pinions. The 94*B*17 steel is satisfactory for pinions up to a certain size, but has not been so successful as a substitute for 4815 in larger sizes (this is based upon performance in laboratory tests and actual field service). Boron has the advantage of increasing the hardenability in the lower levels of the case, where the carbon content is 0·40–0·60%.

The whole problem of correlating the properties of case and core of case-hardened gears with the performance in service is very controversial. Some metallurgists believe that the core properties are of the greatest importance in controlling performance of gears, whilst others feel that core properties have little effect upon the performance of gears having an adequate case depth, except when core properties are associated with the production of undesirable internal stresses during heat-treatment. In general, high core hardness tends to produce tensile stresses at the surface, whereas low core hardness produces compressive stress. This is not an invariable rule, however. Certainly, the performance of case-hardened boron steel gears has illustrated the fact that core hardness, or core hardenability, is not a sole criterion for the selection of carburizing steels.

Case-hardened gears may fail in several ways. Abrasive wear on the surface of the teeth is certainly due to lack of surface hardness.

Breakage at the root of the tooth seems to originate at, or near, the fillet surface, and probably correlates best with the properties of the case. Figure 20 shows the results of dynamometer tests of case-hardened 7-pitch gears. Tests were continued until one or more teeth broke out at the root. It will be noted that the bending fatigue strength of gears made from three types of steel was approximately the same (55,000–80,000 lb./sq. in.), although there was considerable variation in the length of life at a given stress, even with a single steel. This is typical of all fatigue testing, and does not indicate a variation in the properties of the material. In the tractor division of the Harvester Company it is customary to demand that gears carry 125% of the theoretical service for 5 million cycles without failure. The case-hardened boron steel gears that have been used in the tractor division have successfully met this requirement. The bending fatigue strength was as good as that of gears made from other steels.

It seems to be true, however, that bending fatigue strength is affected somewhat by the depth of the hardened layer. If the hardness in the case falls off too rapidly, low results on the dynamometer might occur. Pitting at the pitch line is said to be due either to tensile stresses on the surface or to shear stresses within a very few thousandths of the surface, or to combinations of both. However, the depth of the hardened layer and the hardness of the core adjacent to the case may have an effect upon the resistance to pitting.

More recently, it has been found that the tendency for the case to flake, or spall, off heavy pinions seems to be associated with the depth of the hard layer (the portion of the case which is over 50 R_C). If the hardness in the bottom part of the case is inadequate to carry the load, it is possible to produce shear failures which result in ' shelling off ' the outer portion of the case. In some experimental trials it has been found that the addition of boron to case-hardening steels produces a higher hardness in the bottom portion of the case, even though the surface hardness remains the same. This would be expected, as boron increases the hardenability of steels containing 0·40–0·60% C, but does not appreciably increase the hardenability of steels with 0·90–1·00% C.

On the whole, it may be said that the boron steels have been used satisfactorily to replace other heat-treating steels on the basis of duplicating the hardenability of the original steel. Case hardening, however, involves other problems. Boron steels have been used very satisfactorily for certain gears. In other cases, distortion has been a major difficulty. Before specifying a boron steel for a particular case-hardened gear, the whole subject should be given very thorough study.

Equivalent Dia. of Round Bars Quenched in Mildly Agitated Water, in.					Equivalent Dia. of Round Bars Quenched in Mildly Agitated Oil, in.				
	Surface	¼-Radius	Centre			Surface	¼-Radius	Centre	
3·8					1·8	3·0	3·8		
1·1	2·9				1·0	2·0	2·8	3·6	
0·7	1·6	2·4	3·2	4·0	0·6	1·4	2·1	2·8	3·4

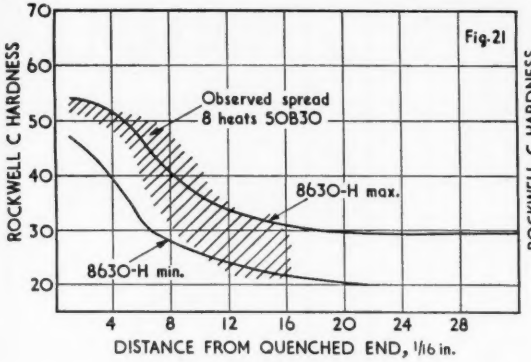
APPENDIX

General Principles for the Application of Boron Steels

As a general rule, boron steels can be used in place of other types of alloy steels for heat-treated parts, if the boron steels are selected on the basis of duplicating the hardenability of the previous steels. Although there may be some exceptions, this rule will work satisfactorily in the great majority of cases. Typical applications are given in Table III, and hardenability charts for various steels and the

Equivalent Dia. of Round Bars Quenched in Mildly Agitated Water, in.

3·8					Surface
1·1	2·9				¾-Radius
0·7	1·6	2·4	3·2	4·0	Centre

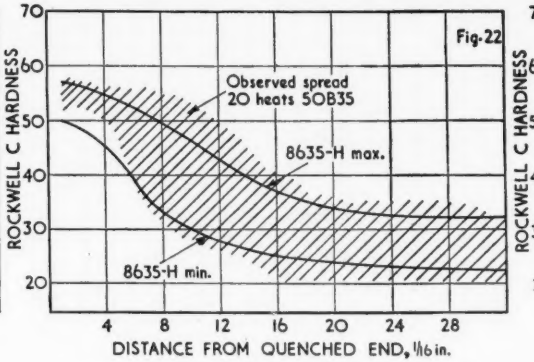


Specified Composition

Steel		C, %	Mn, %	Si, %	Ni, %	Cr, %	Mo, %	B
50*B*30	Min.	0·28	0·75	0·20	...	0·40	...	Yes
	Max.	0·33	1·00	0·35	...	0·60	...	Yes
8630-*H*	Min.	0·27	0·60	0·20	0·35	0·35	0·15	No
	Max.	0·34	0·95	0·35	0·75	0·65	0·25	No

Equivalent Dia. of Round Bars Quenched in Mildly Agitated Oil, in.

1·8	3·0	3·8			Surface
1·0	2·0	2·8	3·6		¾-Radius
0·6	1·4	2·1	2·8	3·4	Centre

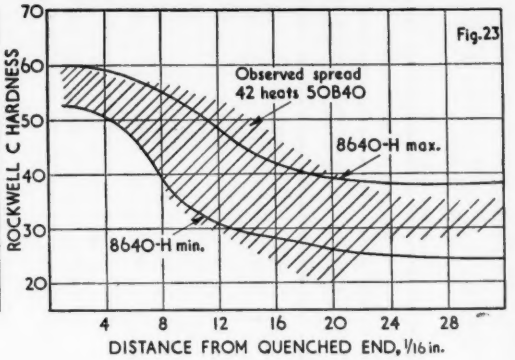


Specified Composition

Steel		C, %	Mn, %	Si, %	Ni, %	Cr, %	Mo, %	B
50*B*35	Min.	0·33	0·75	0·20	...	0·40	...	Yes
	Max.	0·38	1·00	0·35	...	0·60	...	Yes
8635-*H*	Min.	0·32	0·70	0·20	0·35	0·35	0·15	No
	Max.	0·39	1·05	0·35	0·75	0·65	0·25	No

Equivalent Dia. of Round Bars Quenched in Mildly Agitated Oil, in.

1·8	3·0	3·8			Surface
1·0	2·0	2·8	3·6		¾-Radius
0·6	1·4	2·1	2·8	3·4	Centre

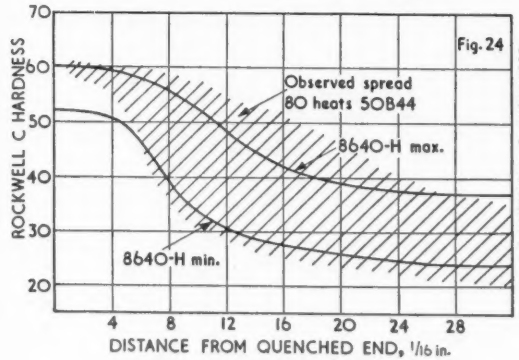


Specified Composition

Steel		C, %	Mn, %	Si, %	Ni, %	Cr, %	Mo, %	B
50*B*40	Min.	0·38	0·75	0·20	...	0·40	...	Yes
	Max.	0·43	1·00	0·35	...	0·60	...	Yes
8640-*H*	Min.	0·37	0·70	0·20	0·35	0·35	0·15	No
	Max.	0·45	1·05	0·35	0·35	0·65	0·25	No

Equivalent Dia. of Round Bars Quenched in Mildly Agitated Oil, in.

1·8	3·0	3·8			Surface
1·0	2·0	2·8	3·6		¾-Radius
0·6	1·4	2·1	2·8	3·4	Centre

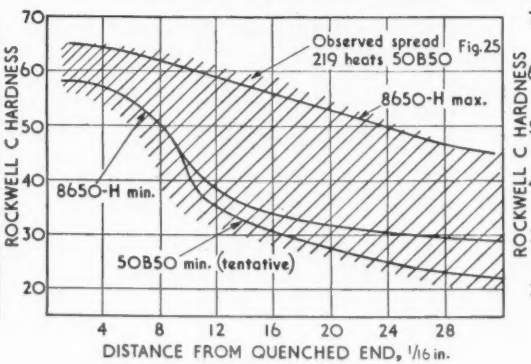


Specified Composition

Stee		C, %	Mn, %	Si, %	Ni, %	Cr, %	Mo, %	B
50*B*44	Min.	0·43	0·75	0·20	...	0·40	...	Yes
	Max.	0·48	1·00	0·35	...	0·60	...	Yes
8640-*H*	Min.	0·37	0·70	0·20	0·35	0·35	0·15	No
	Max.	0·45	1·05	0·35	0·75	0·65	0·25	No

Equivalent Dia. of Round Bars Quenched in Mildly Agitated Oil, in.

1·8	3·0	3·8			Surface
1·0	2·0	2·8	3·6		¾-Radius
0·6	1·4	2·1	2·8	3·4	Centre

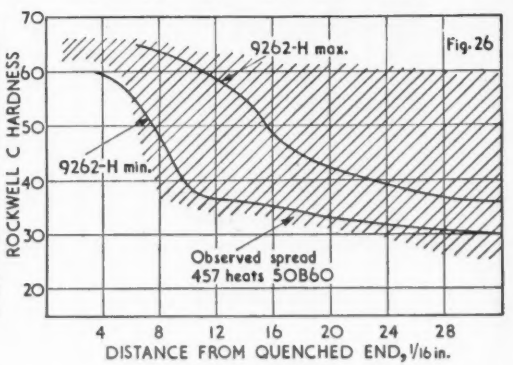


Specified Composition

Steel		C, %	Mn, %	Si, %	Ni, %	Cr, %	Mo, %	B
50*B*50	Min.	0·48	0·75	0·20	...	0·40	...	Yes
	Max.	0·53	1·00	0·35	...	0·60	...	Yes
8650-*H*	Min.	0·46	0·70	0·20	0·35	0·35	0·15	No
	Max.	0·54	1·05	0·35	0·75	0·65	0·25	No

Equivalent Dia. of Round Bars Quenched in Mildly Agitated Oil, in.

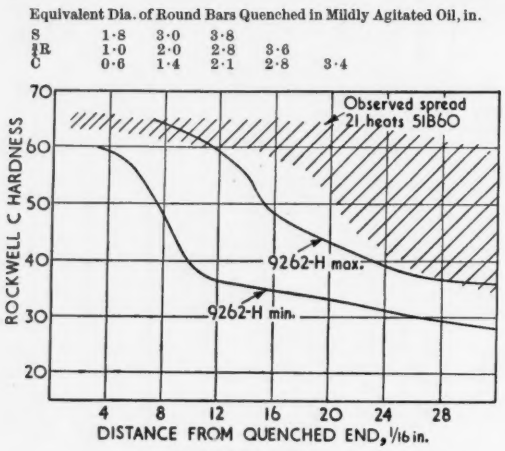
1·8	3·0	3·8			Surface
1·0	2·0	2·8	3·6		¾-Radius
0·6	1·4	2·1	2·8	3·4	Centre



Specified Composition

Steel		C, %	Mn, %	Si, %	Ni, %	Cr, %	Mo, %	B
50*B*60	Min.	0·55	0·75	0·20	...	0·40	...	Yes
	Max.	0·65	1·00	0·35	...	0·60	...	Yes
9262-*H*	Min.	0·55	0·70	1·70	...	0·20	...	No
	Max.	0·65	1·05	2·20	...	0·50	...	No

Figs. 21–26—Hardenability charts prepared from data from the Wisconsin Steelworks of the International Harvester Company



Specified Composition

Steel		C, %	Mn, %	Si, %	Ni, %	Cr, %	Mo, %	B
51*B*60	Min.	0·55	0·75	0·20	...	0·70	...	Yes
	Max.	0·65	1·00	0·35	...	0·90	...	Yes
9262-*H*	Min.	0·55	0·70	1·70	...	0·20	...	No
	Max.	0·65	1·05	2·20	...	0·50	...	No

Fig. 27—Hardenability chart prepared from data from the Wisconsin Steelworks of the International Harvester Company

alternative boron steels are shown in Figs. 21–27. So far as any particular part is concerned, it is only necessary to duplicate that portion of the hardenability band which represents the heat-treatment of that part. For example, consider a 1·6-in. dia. oil-quenched shaft which was previously made of 8640 steel and gave satisfactory performance. Assume that the maximum stress appeared at the bottom of a spline or key at about ¾-radius from the centre of the shaft. This position on an oil-quenched round is represented by a point on the Jominy specimen about $\frac{6}{16}$ in. from the quenched end.

Referring to Fig. 23, the hardenability of 42 heats of 50*B*40, as represented by the Rockwell hardness at $\frac{6}{16}$ in., corresponded exactly with that of the 8640-*H* steel. It would be assumed, therefore, that 50*B*40 could be substituted for 8640 for a shaft of this size. Experience in production has confirmed this. For bolts, or other parts subjected to uniform tensile stresses over the entire area, it might be desirable to limit the application of 50*B*40 to slightly smaller sizes.

Figure 23 shows also that the minimum hardness which would be expected at ¾-radius on a 4-in. dia. oil-quenched shaft is somewhat lower than that which would be produced with 8640-*H* steel. This is represented by the $\frac{17}{16}$ in. point on the Jominy bar. In practice, however, neither of these steels would be used for rounds of this size, because the hardness produced is too low to be of practical value.

Another useful procedure which is sometimes followed is to select steel which will produce 90% martensite with a commercial quench at the locality of highest stress; *e.g.*, at ¾-radius. The hardness corresponding to 90% martensite, or 0·40% carbon alloy steels, is approx. 50 R_C. Table IV shows the minimum Rockwell hardness that would be expected at ¾-radius, after a commercial oil-quench with different steels and bars of different diameter. Note that 8640 steel meets this requirement only up to about 1⅛-in. dia., whereas it is frequently used for heat-treated parts up to about 1½–1¾ in. dia. The 50*B*40 steel is comparable with 8640, and 86*B*45 is comparable with 4340.

Another criterion which is sometimes used is that the as-quenched hardness should be at least 10 R_C higher than the specifications for minimum hardness

Table IV

EXPECTED MINIMUM AS-QUENCHED HARDNESS AT ¾-RADIUS

(Quenched in oil)

Steel	Rockwell C Hardness at:		
	1-in. dia. $\frac{4}{16}$ Jominy	2-in. dia. $\frac{8}{16}$ Jominy	3-in. dia. $\frac{12}{16}$ Jominy
4340	52	52	50
86B45	54	52	49
4140	52	47	38
81B40	52	50	38
4640	50	37	27
8640	51	39	30
50B40	51	40	27
14B50	46	28	25

Table III

TYPICAL APPLICATIONS

Part	Size	Brinell Hardness	Quench	Original Steel Specification	Alternative Boron Steel
Shafts	> 3 in	Up to 450	Oil	4340	86B45
Flat springs	1–2 in. thick	Up to 450	Oil	9262	51B60
Coil springs	1–3 in. dia.	Up to 450		8660	51B60
				9260	50B60
Flat and coil springs	< 1 in. dia.			5152	80B60
Shafts, forgings, heat-treated parts	< 2 in. dia.	Up to 450	Oil	4140	81B40
	3 in. dia.		Water		
	> 3 in. dia.		Water	8640	80B40 or 50B40
Bolts	< $\frac{9}{16}$ in.		Oil	4130	
	1–1¼ in.		Water	4035	14B35
				3130	
Bolts requiring difficult cold heading			Oil	4037	40B37

Table V

SPECIFIC APPLICATIONS OF HEAT-TREATING BORON STEELS

Type of Part	Critical Section	Heat-Treatment	Hardness		Original Steel Specification	Alternative Boron Steel	Remarks
			Surface	Centre			
Axles and Shafts							
Jackshaft	$1\frac{3}{4}$-in. O.D. spline	O.Q., temper	331–388 Brinell	...	4140-H	TS86B45	All of these parts have hardened as well out of the boron steels as out of the specified steels. Little change in required tempering temperature. All parts are machined to finished size before heat-treatment, except for grinding stock and hobbing splines. The hardening furnaces are gas-fired with no controlled atmosphere. No unusual trouble experienced in hobbing at the maximum hardness. Axle splines are shot-peened. A few torsion tests have shown torsional strengths similar to those obtained when using the specified boron-free steels
Propeller shaft	$1\frac{3}{4}$-in. O.D. spline	O.Q., temper	321–388 Brinell	...	4340-H	81B45	
Transfer shaft	$1\frac{3}{4}$-in. O.D. spline $2\frac{1}{2}$-in. O.D. spline	O.Q., temper	341–401 Brinell	...	4340-H	TS86B45	
Short axle	$2\frac{17}{32}$-in. O.D. spline	O.Q., temper	401–461 Brinell	375 min. Brinell	4340-H	TS86B45	
Long axle	$2\frac{17}{32}$-in. O.D. spline	O.Q., temper	401–461 Brinell	375 min. Brinell	4340-H	TS86B45	
Medium axle	$2\frac{23}{64}$-in. O.D. spline	O.Q., temper	401–461 Brinell	375 min. Brinell	4340-H	TS86B45	
Automatic transmission main shaft	1-in. O.D. splined end with $\frac{1}{4}$-in. bore	O.Q., draw bar stock to 19–24 R_C. Then cold forge, which raises critical section to 28–32 R_C			8640	80B40	Satisfactory in torsion fatigue. Used as alternative, but not often as yet. Bearing diameters induction-hardened to 58–62 R_C. Draw 163° C.
Axle shaft	$2\frac{1}{2}$-in. rad. Root of spline $\frac{1}{16}$ in. below surface	W.Q., draw	38–44 R_C	36 R_C min. at root of spline	8641	50B44	Must maintain minimum of 8645-H hardenability at $\frac{1}{4}$ Jominy to produce required hardness in root of spline
Axle shaft	2-in. dia. Root of spline $\frac{1}{16}$ in. below surface	Induction-hardened, O.Q.	40–47 R_C	40–47 R_C at root of spline	8641	50B44	
Through shafts	2-in. dia.	O.Q., temper	375–430 Brinell	...	4340 3140	94B40	Equal to original
Axle shaft	1·3-in. dia. splined section	Caustic quench, draw	331 Brinell	302–321 Brinell	1330 8630	14B38	Comparable to original steel. Approx. 116 tons used to date on this part
Shaft	2·400-in. P.D. splined section	O.Q., draw	32–36 R_C		4340-H	TS86B45	
Axle shaft	$1\frac{1}{2}$-in. dia. splined section	O.Q., draw	363–444 Brinell	330 min. Brinell	8645	80B50	Torsion fatigue strength as good as original
Fuel pump main shaft	$1\frac{1}{4}$-in. dia.	O.Q. 871° C., draw	Full cross-section: 28–35 R_C		4140-H	80B50	Core hardness satisfactory, but much more warpage, owing to high quenching temperature required
Axle centres	Large I beam sections, 3–4 in.	W.Q., temper	285–341 Brinell	...	4130 3130	94B30	Equal to original
Trailer axles, I beam sections	4-in. max.	O.Q., temper	269–321 Brinell	...	4140	94B40	Equal to original
Axle shaft: 95030	1·285-in. spline, 1·563-in. taper	O.Q., draw	321–388 Brinell	331–363 Brinell	8650-H	50B50	
86778-R1	1·515-in. spline $\frac{3}{8}$ in. × 5·352 in. flange	O.Q., draw	321–388 Brinell 229–321 Brinell	363–375 Brinell	8650-H	50B50	No difference
69187	1·750-in. spline $\frac{3}{8}$ in. × 5·75 in. flange	O.Q., draw	321–388 Brinell 229–321 Brinell	340–363 Brinell	8650-H	50B50	
68932-R1	1·882-in. spline $\frac{7}{16}$ in. × 6·25 in. flange	O.Q., draw	321–388 Brinell 229–321 Brinell	363–375 Brinell	8650-H	50B50	
Tractor axle	3-in. splined	O.Q., draw 454° C. approx.	321–388 Brinell		8645	50B50 and sulphite	
Rear axle	$2\frac{3}{4}$-in. fillet	W.Q., draw	321–388 Brinell		4145 8641	50B50	Physical properties good. Slight tendency to cracking. 50B44 being tried
Front axle	$1\frac{3}{8}$-in. fillet	W.Q., draw	293–352 Brinell		8650	50B50	" " " "
Power take-off shaft	$1\frac{3}{8}$-in. dia.	O.Q., draw 538° C. approx.	285–341 Brinell		8640	50B50 and sulphite	

Table V—(*Continued*)

SPECIFIC APPLICATIONS OF HEAT-TREATING BORON STEELS

Type of Part	Critical Section	Heat-Treatment	Hardness		Original Steel Specification	Alternative Boron Steel	Remarks
			Surface	Centre			
Shaft-centre bearing: 62548-R1	2½-in. dia.	O.Q., draw	365–444 Brinell		8650-H	50B50	
Steering clutch shaft coupling	Hobbed splined section	O.Q., draw	34–41 R_C		8640-H	50B44	
Track roller shaft	Milled section	O.Q., draw	35–42 R_C		8640-H	50B50	
Steering clutch shaft	3½-in. dia. splined section	O.Q., draw	34–44 R_C		8640-H	50B50	
Bolts, Studs, and Nuts							
Nut	3⅛-in. O.D., 1⅛-in. 16 N2 thread I.D.	O.Q., draw	26–32 R_C		8740	TS81B40	Used in production; no special tests made
Stud	⅜-in. threads	Heat-treated bar stock	26–32 R_C		8740 4140 8640	TS81B40	,, ,, ,, ,,
Stud	½-in. threads	Heat-treated bar stock	26–32 R_C		8740 4140 8640	TS81B40	,, ,, ,, ,,
Bolts	½ in.	Quench, draw			4140	94B40	130,000 min. yield
Cap screws ⩽ ½ in.			286–321 Brinell		8637	80B40	Satisfactory
Connecting rod bolts			286–321 Brinell		8637	80B40	Satisfactory
Cylindrical head stud	⅞-in. dia.	O.Q., draw	Full cross-section: 30–35 R_C		4140-H	81B45	Tensile properties above minimum limits for 4140-H
Large bolts and studs	1-in. dia.	O.Q., temper	269–321 Brinell	...	3135 4140	94B40	Equal to original
Nut, arm bearing retaining	1⅛ in.	Quench, draw	30–35 R_C		8640	81B40	Equivalent
Connecting Rods and Crankshafts							
Military tank crankshaft			286–321 Brinell. Nitrided case 0·015–0·025 in.		4340	86B45	Satisfactory. Type test work. Completed 500 hr.
Connecting rods			269–302 Brinell		8640	80B40	Satisfactory. Type test work. Completed 500 hr.
Connecting rods: 62930-R1	I sections, ⅛ in. × 7/16 in. and ⅛ in. × ⅜ in.	O.Q., draw	217–269 Brinell	217–269 Brinell	8632-H	50B30	
118701-H	⅛ in. × ⅜ in. and 1/16 in. × 1½ in.	O.Q., draw	217–269 Brinell	217–269 Brinell	8632-H	50B30	
Miscellaneous Forgings							
All stock ⩽ 7/16 in. dia. round			286–321 Brinell		8640	80B40	Satisfactory
Equalizer spring clip	Threaded section ½-in.	O.Q. draw	28–34 R_C		8640-H	50B50	
Track spring adjusting rod	Threaded section ¾-in.	O.Q., draw	23–31 R_C		8640-H	50B44	
Bumper stop	⅝-in. flat	O.Q., draw	321–388 Brinell		8640	50B40	
Anchor, torsion bar	¾-in. splined section	Quench, draw	32–36 R_C		8640	81B50	Equivalent
Track shoe	¾ in.	Quench, draw	28–32 R_C		8740	50B40	Equivalent
Torsion bar	1·9-in. dia.	Quench, draw	46–50 R_C		8660	81B60	Torsional fatigue strength appears to be as good as original material
General machine tool parts, shafts, gears, etc.	½–3 in.	O.Q., draw	29–33 R_C 44–47 R_C		8650	81B45	Very good results. Have used steel from over 50 heats
	3–6 in.	O.Q., draw	29–33 R_C 44–47 R_C		4340	86B45	Very good results. Have used steel from over 16 heats
Medium sized injector body	1½-in. max. cross-section	O.Q., draw	Min. hardness at centre of section, 30 R_G		4340	81B50	Heat-treating characteristics satisfactory

Table V—(*Continued*)

SPECIFIC APPLICATIONS OF HEAT-TREATING BORON STEELS

Type of Part	Critical Section	Heat-Treatment	Hardness		Original Steel Specification	Alternative Boron Steel	Remarks
			Surface	Centre			
Large-sized injector body	2¼-in. cross-section	O.Q., draw	Min. hardness at centre of section, 30 R_G		4340	81B50	Plunger hole drilling and honing satisfactory. No harmful amount of free ferrite in core
Intermediate road wheel arm	3½-in. dia.	Quench, draw	311–341 Brinell		4350	86B50 (J44 at 22/16)	Equivalent
Sprocket drive pinion	Ring grove	Rough stock. Harden, W.Q.	30–37 R_c		8645	81B45	Trouble with machining
		Induction-harden, O.Q.	55 R_c min.				
Steering Parts							
Steering knuckle		O.Q., draw	248–321 Brinell	248 Brinell	8640	80B40	No trouble
Steering arms	1–2-in. dia.	W.Q., temper	285–341 Brinell		3130 4130	94B30	Equal to original
Steering knuckles	1–4-in. dia.	W.Q.	285–341 Brinell		3130 4130	94B30	Equal to original
Spindles	Tubular section, ⅛-in. wall	W.Q., temper	269–321 Brinell		4130 3130	94B30	Equal to original
Spindles	Tubular section, ⅛-in. wall	O.Q., temper	285–341 Brinell		3140	94B40	Equal to original
Steering arm: 69589-R1	¾-in. dia. to 1½ in. × 1 in. square	O.Q., draw	269–321 Brinell	277 Brinell	8640-H	50B40	
85053-R1	⅝ in. dia. to 1 in. × ⅞ in. square	O.Q., draw	269–321 Brinell	269–285 Brinell	8640-H	50B40	
100771-H	1/16 in. × 1 in. to 1/16 in. × 1½ in. square	O.Q., draw	269–321 Brinell	255–285 Brinell	8640-H	50B40	
95105-H	⅝-in. dia. to 1 in. × 1⅛ in. square	O.Q., draw	269–321 Brinell	255–285 Brinell	8640-H	50B40	
Steering knuckle: 69567-R1	1·350-in. dia. spindle	W.Q., draw	285–341 Brinell	235–302 Brinell	8630-H	50B30	
69592-R1	1·562-in. dia. spindle	W.Q., draw	285–341 Brinell	235–302 Brinell	8635-H	50B35	
Steering arm	Boss 1½ in. × 1½ in. × 1½ in.	O.Q., draw	30–37 R_c		New part	50B40	
Front axle pivot shaft	1½-in. dia.	O.Q., draw	321–388 Brinell		8640	50B40	
Steering knuckle	1½-in. dia.	O.Q., draw approx. 468° C.	321–388 Brinell		8645	50B50 and sulphite	
110099-R1	2⅛-in. dia. spindle	O.Q., draw	285–341 Brinell	255–285 Brinell	8645-H	50B44	
Shackle	2-in. dia.	Quench, draw	285–311 Brinell		8745	81B55	Equivalent
Spindle	3¼-in. dia.	Quench, draw	35–39 R_c		4340-H	86B50 (J50 at 22/16)	Equivalent
Springs							
Passenger car leaf springs	1¾ in. × 0·214 in. 0·237 in. 0·262 in. 0·291 in. 2 in. × 0·214 in. 0·237 in.	O.Q., draw	418–430 Brinell	Same	A.I.S.I. 5155	50B60	About 570 tons used
Truck and tractor leaf springs	Up to 1 in. thick	O.Q., draw	375–444 Brinell	Same	9260	50B60	450 heats
Tractor springs	Over 1 in. thick	O.Q., draw	375–444 Brinell	Same	9262	51B60	21 heats

Table VI

SPECIFIC APPLICATIONS OF CASE-HARDENING BORON STEELS

Type of Gear or Pinion	Pitch of Teeth	Maximum Thickness Under Teeth	Case Depth	Treatment	Rockwell Hardness		Original Steel Specification	Alternative Boron Steel	Remarks
					Case	Core			
Transmission Gears and Pinions									
Automatic transmission ring gear P.D. 4·58	15 D.P.	$\frac{1}{4}$ in.	0·015–0·025 in.	Gas - carburize, Gleason press quench, draw 163° C.	15N 89–91	30–40	8620	80B20	Satisfactory. Used as alternative, but not often
Automatic transmission sun gear with integral free-wheel race	15 D.P.	$\frac{1}{16}$ in.	0·015–0·020 in.	Gas - carburize, direct quench, draw 204° C., then induction-harden free-wheel race, 0·060–0·090 in. deep. Draw 177° C.	15N 88–91 C 60–62		8640	80B40	Satisfactory. Some difference in dimensional change in heat-treatment. Performance of 80B20 not good on larger heavy duty gears, 8 pitch
Transmission spin gear	7	$1\frac{23}{64}$ in.	0·035–0·045 in.	Direct O.Q., low draw	57–63 R_c		4120 8620 4118	50B15	Satisfactory, except some distortion
Transmission Gears:									
Mainshaft 44073-H	7–9	$\frac{5}{8}$ in.	0·031–0·047 in.	Marquench, draw	62–64	27–34	4620-H	50B20	Where core hardness is in the same hardenability band as the 8600 series steel it replaces, the distortion is comparable
90850-HC	7–9	$1\frac{1}{8}$ in.	0·025–0·035 in.	Marquench	62–66	28–37	8620-H	50B15	
Main drive 80021-R1	9·25	$\frac{1}{2}$ in.	0·025–0·035 in.	Marquench, draw	58–60	26–35	8620-H	50B15	
74899-H	7–9	$\frac{3}{4}$ in.	0·031–0·047 in.	Marquench, draw	60–64		4620-H	50B15	
Countershaft 90849-H	7–9	$1\frac{1}{4}$ in.	0·025–0·035 in.	Marquench, draw	62–64	31–45	8620-H	50B20	
44057-H	7–9	2 in.	0·031–0·047 in.	Marquench, draw	61–63		4620-H	50B15	Core hardness readings are $\frac{1}{8}$ in. below root of tooth
Mainshaft 80119-R2	(Splines only)		0·025–0·035 in.	Marquench, draw	60–65	23–25	8620-H	50B15	Where core hardness is in the same hardenability band as the 8600 steel it replaces, the distortion is comparable
82220-HA	(Splines only)		0·040–0·055 in.	Marquench, draw	63–65	30–34	4620-H	51B20	
Countershaft 51615-H	Plain shaft		0·031–0·047 in.	Marquench, draw	63–65	23–27	8620-H	51B20	Core hardness readings are $\frac{1}{8}$ in. below root of tooth
63433-H	6–7	$1\frac{1}{8}$ in.	0·040–0·055 in.	Marquench, draw	62–64	12–20	4820-H	51B20	
Helical gear	6	0·6 in.	0·045–0·055 in.	Carburize, slow cool, reheat, quench, draw	60 min.	30 min.	3312 9310-H	TS43BV14	Used in production; no special tests made
Heavy section tractor transmission gears	6	2 in.	0·040–0·050 in. 0·050–0·060 in.	Carburize, slow cool, reheat, quench, draw	60 min.	30–45	8620	50B20	Satisfactory on heavy sections. Distortion a problem on light sections
Heavy Duty Gears—Bevel and Hypoid									
Straight-tooth bevel pinion and shaft	4	$1\frac{3}{4}$ in.	0·050–0·060 in.	Direct quench from 884° C., draw 657° C., reheat 871° C. Gear end in lead. O.Q., draw 162° C.	57–63 and 28–37	32–48	8620	50B20	Problem of hardness in root until quench was revised
Bevel ring gear, straight tooth	4	$1\frac{1}{2}$ in.	0·050–0·060 in.	Direct quench from 871° C., draw 162° C.	57–63	30–45	8620	50B20	
Straight-tooth bevel pinion for thresher	$4\frac{1}{2}$	$\frac{1}{2}$ in.	0·040–0·050 in.	Direct O.Q. from 884° C., draw 162° C.	57–62		8620	50B15	Satisfactory
Differential side gear	$2\frac{1}{2}$–4	1 in.	0·050–0·060 in.	Direct quench, temper	57–63	23–35	4620	94B17	Equal to original
Truck and tractor driving ring gear	$3\frac{1}{2}$	1·2 in.	0·072 in.	Direct quench, draw	55	34	8622 4617	94B17	Small heat tested and not replaced on production basis
Truck and tractor rear axle hypoid gear	5	$2\frac{13}{64}$ in.	0·072 in.	Direct quench, draw	58–60	23	8620 4617	94B17	Small heat tested and not replaced on production basis. Satisfactory for hardenability and distortion without any major changes in processing
Hypoid pinion	2–4	5 in.	0·070–0·090 in.	Direct quench, temper	57–63	25–35	4620	94B17	Equal to or better than original

Table VI—(*Continued*)

SPECIFIC APPLICATIONS OF CASE-HARDENING BORON STEELS

Type of Gear or Pinion	Pitch of Teeth	Maximum Thickness Under Teeth	Case Depth	Treatment	Rockwell Hardness		Original Steel Specification	Alternative Boron Steel	Remarks
					Case	Core			
Hypoid gear	2–4	2½ in.	0·050–0·070 in.	Direct quench, temper	57–63	25–35	4620	94B17	
Final drive pinion	4	1½ in.	0·050–0·070 in.	Quench, draw	58–62	25–35	4820	43BV14	Equivalent
Spiral bevel gear	3·1416	1·5 in.	0·050–0·060 in.	Carburize, slow cool, reheat, quench, draw	60 min.	30 min.	3312 3310-H	TS43BV14	Equivalent
Special bevel pinion	2·0	2½ in.	0·060–0·080 in.	Carburize, box cool, reheat, press quench	75–85 Sclero. Spec. 56–62 R_C		2515	43BV14	Diesel locomotive application. Some trouble experienced with scoffing. Gear 25-in. dia.
Special bevel gear		2 in.	0·060–0·080 in.	Carburize, box cool, reheat, press quench	56–62 R_C		2515	43BV14	See above
Special bevel gear pinion	2·3	2½–4½ in. dia.	0·060–0·075 in. 0·060–0·080 in.	Box cool, reheat, press quench, gas carburize, direct quench	57–63 57–63 File hard		4320 4320	43BV14 94B17 or 43BV14	21-in. gear for heavy duty truck drive. 94B17 gives distortion in pinion as excessive change
Helical bull gear	3·36	2–4½ in. dia.	0·055–0·065 in.	Carburize, direct quench on press draw	57–65		4620-H	94B17	Double reduction 6 × 6
Helical pinion	3·36		0·055–0·065 in.	Carburize, direct quench, draw	57–65 File hard		4620-H	94B17	Truck axle. Working satisfactorily
Spur bull gear	3·0	2½–4½ in. dia.	0·060–0·075 in.	Box cool, reheat, press quench, draw	55–62		4320-H	43BV14	Satisfactory. Power grader
Spur bull pinion	3·0		0·050–0·070 in.	Carburize, direct quench	57–63 File hard		4820-H	43BV14	
Spiral bevel cluster with 2 sets of spurs	5·0	1–2½ in. dia.	0·040–0·050 in.	Carburize, direct quench cap to keep bores soft for size broaching	57–63 File hard		4820-H	94B17	Tractor drive set
Special bevel pinion with 1 set of spurs	5·0		0·040–0·050 in.	Carburize, direct quench	57–63 File hard		4820	94B17	
Special bevel pinon	1·8	4½ in. dia.	0·060–0·070 in.	Gas-carburize, direct quench	57–63 File hard		3310	43BV14	Drive pinion—earth mover

Type of Part	Critical Section	Case Depth	Treatment	Rockwell Hardness		Original Steel Specification	Alternative Boron Steel	Remarks
				Case	Core			
Case-Hardened Bearing Races								
(No changes or contemplated changes to boron steels. Information below a matter of checking only)								
Bearing inner race—tapered	⅛–¼ in. wall	0·060–0·070 in.	Direct quench, reheat, O.Q., draw	60–64		Original: 4620 Present: 4720 8620	94B17	Structure and hardness satisfactory. Service performance being watched
Bearing outer race—tapered	⅛–$\frac{7}{32}$ in. wall	0·055–0·065 in.	Cool, reheat, O.Q., draw	60–64		Original: 4608 Present: 4012	14B20	Not satisfactory for oil hardening unless Mn content is about 1·00%
14B20: Not considered. Will not oil harden 94B17: Not considered. Can see no advantage for a bearing race, compared with 8620, for case hardness, structure, economy, or nickel conservation								
Small automotive-type bearing cone 09067	Max. thickness	0·040–0·050 in.	O.Q., reheat, O.Q., draw	61–62	25–27	4620 4720	80B20	Machinability slightly better than standard analysis. Increased final hardening temperature improved core hardness and properties. Case and core structure very satisfactory. At last report some 50,000 miles in front wheel assembly of company cars (3)

Table VI—(*Continued*)

SPECIFIC APPLICATIONS OF CASE-HARDENING BORON STEELS

Type of Part	Critical Section	Case Depth	Treatment	Rockwell Hardness		Original Steel Specification	Alternative Boron Steel	Remarks
				Case	Case			
Small automotive-type bearing cone 09067	Max thickness	0·040–0·050 in.	O.Q., reheat, O.Q., draw	61–62	25–27	4620 4720	80B20	To date, rolls only have been heat-treated. Hardening temperature of 801° C. and above gave satisfactory properties (815° C.). When complete, this series is to be tested in lab., and also in field on customer equipment
Highway equipment (truck and trailer)								
745 Series:								
749 Cone	$\frac{1}{32}$-in. max. thickness	0·060–0·070 in.	Same as above			4720	94B17	
742 Cup	$\frac{1}{8}$-in. max. thickness	0·060–0·070 in.				4720	94B17	
702 Roll	$\frac{3}{4}$-in. rad.	0·070–0·080 in.		62–63	30–32	4620	94B17	
655 Series:								
663 Cone	$\frac{1}{32}$-in. max. thickness	0·065–0·075 in.				4720	94B17	
672 Cup	$\frac{1}{8}$-in. max. thickness	0·070–0·080 in.	O.Q., reheat, O.Q., draw			4720	94B17	655 and 745 series are being processed simultaneously with the rolls being the only parts thus far complete. Results of the 602 were comparable to the 702
602 Roll	$\frac{3}{4}$-in. rad.	0·065–0·075 in.		61–62	30–32	4620	94B17	
Miscellaneous Case-Hardened Parts								
Differential spiders	1-in. dia.	0·045–0·060 in.	Direct quench	57–63	25–40	4620 2512	94B17	
Straight knuckle pins	1-in. dia., approx.	0·050–0·070 in.	Direct quench	57–63	25–40	2512 4620	94B17	
Straight ball studs	1-in. dia., approx.	0·050–0·070 in. on ball only	Direct quench	57–63 on ball	25–40	3120 4620	94B17	
Splined clutch shaft	$1\frac{1}{2}$-in. dia., 10 rectangular splines	0·025–0·035 in.	Carburize, cool, reheat, O.Q., temper	56–62		4620-H	94B17	Carburizing rates, attained hardness, etc., same for boron steels.
Pivot shaft	2-in. O.D. round pin	0·030–0·040 in.	Carburize, cool, reheat, O.Q., temper	58–63		4620-H	94B20	Little data on distortion, as these are straight shafts which never give serious distortion problems
Splined clutch shaft	$1\frac{1}{2}$-in. dia., 10 rectangular splines	0·025–0·035 in.	Carburize, cool, reheat, O.Q., temper	56–62		4620-H	94B17	
Governor shaft, $7\frac{1}{2}$-in. long slender part	$\frac{5}{8}$-in. dia. max. down to 0·330 in. dia. min., hole drilled through most of length of part	0·030–0·040 in. effective case, 0·040–0·050 in. total case	Cool, reheat, O.Q., temper	60–63 R_c	25 R_c	8620-H	80B20	Core hardness and strength satisfactory
Idler shaft, fuel pump	$\frac{5}{8}$-in. dia.	0·015–0·025 in. effective case, 0·030-in. max. total case	Cool, reheat, O.Q., temper	79–81 R_a		8620-H	80B20	Entirely satisfactory
Universal joint spider	$\frac{3}{4}$-in. dia.	0·032–0·040 in.	Direct quench, draw	61–62	41–42	8620	80B20	Comparable to original steel
Truck and tractor differential spider	1-in. dia.	0·032–0·040 in.	Direct quench, draw	60–61	36–37	8620	80B20	Comparable to original steel
Ball, shackle bearing	1 in.	0·025–0·030 in.	Quench, draw	60		8620	80B20	Equivalent
Steering knuckle pin	$1\frac{5}{8}$ in.	0·047–0·063 in.	Cool, reheat, W.Q., draw	57–63		4120 8620 4118	50B15	Physical properties are satisfactory
Steering knuckle pin	$1\frac{57}{64}$ in.	0·060–0·075 in.	" "	"		"	"	" " "
Piston pin	$\frac{7}{8}$-in. O.D., $\frac{3}{8}$-in. I.D.	0·045–0·055 in.	Cool, reheat, O.Q.	60	30–35	C1118	80B20	No trouble, small lot used
Piston pins	0·890–0·896 in. dia. stock			58–60 R_c	26–35 R_c	1020	40B37	Satisfactory
Camshafts (military tank engine)				58–62 R_c	26–35 R_c	8620	80B20	Satisfactory. Completed 500 hr.

after tempering. For example, consider a specification of 321–388 Brinell, which corresponds to 34 R_C minimum. If an as-quenched hardness of 10 R_C above the final hardness is specified, this would mean an as-quenched hardness of 44 R_C, which would permit the use of 50*B*40 and 8640-*H* for oil-quenched sections up to about $1\frac{3}{4}$ in. dia.

Details of specific applications of boron steels in use in 19 American companies are given in Tables V and VI.

Bibliography

1. E. E. THUM (Editor): "Boron Steel," 2nd revised ed.: 1953, Cleveland, Ohio, American Society for Metals.
2. F. X. KAYSER, R. F. THOMSON, and A. L. BOEGEHOLD: *Trans. Amer. Soc. Metals*, 1953, vol. 45, pp. 1056–1079.
3. R. A. GRANGE, W. B. SEENS, W. S. HOLT, and T. M. GARVEY: *Ibid.*, 1950, vol. 42, pp. 75–111.
4. C. F. JATCZAK and E. S. ROWLAND: *Ibid.*, 1953, vol. 45, pp. 771–786.
5. R. A. GRANGE and J. B. MITCHELL: "Effect of Carbon and Boron on the Hardenability of a Case-Carburized Steel." (Paper presented at Western Metal Congress, American Society for Metals, in Los Angeles on 24th March, 1953.) United States Steel Corporation Research Laboratory, New Jersey.
6. SOCIETY OF AUTOMOTIVE ENGINEERS: "S.A.E. Reports on Boron Steels" (SP–81).
7. H. B. KNOWLTON: "Boron and Other Low-Alloy Steels," S.A.E. Preprint No. 700.
8. A. L. BOEGEHOLD: *S.A.E.J.*, 1950, vol. 58, Jan., pp. 47–51.
9. D. I. BROWN: *Iron Age*, 1951, vol. 168, July 5th, pp. 79–85; 12th, pp. 85–90; 19th, pp. 102–106; 26th, pp. 68–72; Aug. 9th, pp. 75–80.
10. W. G. WILSON: *Foundry*, 1951, vol. 79, Sept., pp. 88–89, 164–165.
11. A. I. KRYNITSKY and H. STERN: *Trans. Amer. Found. Soc.*, 1949, vol. 57, pp. 375–488.
12. R. A. DYKE, jun., and C. K. DONOHO: *Ibid.*, 1952, vol. 60, pp. 797–802.
13. H. KRAINER and M. KRONEIS: *Stahl u. Eisen*, 1952, vol. 72, pp. 1652–1657
14. P. R. WRAY: Technical Paper No. 123: New York, United States Steel Corporation.
15. T. G. DIGGES: Unpublished work.
16. "Boron Treated Cast Steels": Cleveland, Ohio, Steel Founders' Society of America.
17. A. HURLICH: *Ordnance Magazine*, 1953, vol. 37, No. 196, Sec. 1, p. 696.
18. G. W. POWELL, M. B. BEVER, and C. F. FLOE: "The Carbonitriding of Boron Steels." Massachusetts Institute of Technology, Cambridge, Mass. (Based on thesis of G. W. Powell, M.I.T., 1952, Sept.).
19. H. B. KNOWLTON, F. SAILER, and E. H. SNYDER: *Metal Progress*, 1943, vol. 44, pp. 1104–1111.
20. JOINT A.I.S.I.–S.A.E. HARDENABILITY DATA COMMITTEE: Hardenability Data Report, 1944, May 19th.

DISCUSSION

After Mr. **H. B. Knowlton** had read his paper, Mr. **R. Wilcock** presented his paper entitled "*The Effect of Boron on the Mechanical Properties of Low-Alloy Steels*," which appeared in the *Journal* for April, 1953, vol. 173, pp. 406–419.

Mr. **Brinley Jones** (Climax Molybdenum Co., Ltd.): Mr. Knowlton's paper was most instructive, and I am only sorry that we have not been able to study it in advance and thus be able to express more articulate appreciation of its contents. I should also like to congratulate Mr. Wilcock on the precision with which he has presented his results. For the range of steels concerned he has explained the conditions in which the use of boron is technically sound, and also the conditions in which it does harm (by which I mean specifically the destruction, or part destruction, of the toughness developed in boron-free steels).

In endeavouring to sift and assess the potentialities for exploitation of medium-carbon alloy heat-treated steels which can be inferred from Mr. Wilcock's paper, we are drawn immediately into a consideration of two widely differing principles of steel production and application, one usually associated with U.S. practices and the other more broadly instanced in British and European conditions.

To take first the one extreme of mass production, where the objective lies in the manufacture of huge numbers of identical parts of small or medium section for known service applications, and where it is possible to draw up specialist compositions and apply a rigorously controlled quench to individual components, it seems quite certain that, in principle, boron can be made to substitute on a very wide scale for the more conventionally used alloys. In such conditions, steel composition, heat-treatment, and service performance can be closely correlated. This procedure must, of course, be carried out on a reasonably large scale if it is to pay.

In this case, successful performance is the only criterion. The interposition of a notched-bar test, or any other mechanical test, is not only superfluous but irrelevant. The whole position is implied in a single sentence of Mr. Wilcock's: "In certain sizes the addition of boron makes it possible to achieve a tensile strength unobtainable in the same steel without boron, except by tempering at temperatures lower than those considered suitable."

It is quite sound, in these circumstances, to take risks and, so long as extra strength is being obtained, to continue with boron steel even for components too large to be hardened right through. There is a whole range of applications in which the main property required is fatigue strength, especially in bending or torsion, where boron can be made to serve. I believe there is general recognition and appreciation of the immense efforts which the Americans have made in developing this knowledge.

Substitution as such, however, represents only part of the picture. Much depends on whether one is driven by necessity to use boron. If not, the advantages are not nearly so attractive. Many other points must also be considered. It might be generally interesting, for example, if Mr. Wilcock could give us some information on melting costs; to what extent do we gain or lose by substituting boron? Another problem seems to be the difficulty, with ultra-low alloy steels of the boron-containing type, of obtaining analysis ranges sufficiently narrow to ensure consistent properties. For example, the unpredictability of the distortion occurring during heat-treatment is reported to represent a major complication, particularly with the carburizing steels; after several years of experience with this type, one U.S. authority[1] has recently stated that "The excessive movement encountered in processing carburized gears is, in our opinion, still the most distressing problem to be

overcome." Again, the prevention of wide variations in core strength for small variations in section size would appear to represent a very real problem for case-carburized parts generally.

Can Mr. Knowlton explain why some organizations in the United States placed ideally for using boron steels have not, in fact, adopted them ?

Taking now the opposite extreme from mass production, the situation in this country is that there is a big production of alloy steels, notably heat-treated bars and sections for subsequent machining into miscellaneous parts. These are produced not for some definite application, because the application may not be known to the steelmaker, but to conform to some standard mechanical-properties specification, which includes a notched-bar requirement. It is true that, for many applications, service stresses may be such as not to justify the imposition of a toughness index of this kind. In others, however, conditions of stress concentration and the danger of brittle fracture will undoubtedly occur. Again, the steel supplier will not be apprised of the depth of machining applied in the production of many finished components and the consequent exposure of the interiors of bars to combined stress. There seems little choice, consequently, but to apply generally some form of toughness figure as a safeguard. In these circumstances, it seems fatal that boron should cause such drastic deterioration of toughness in conditions where incomplete hardening is concerned. I believe it would be an impossibility, without serious curtailment of output, to think in terms of full hardening throughout the section of a large proportion of the bar material thus supplied.

Possibly only negligible quantities receive individual quenching treatment, as in Mr. Wilcock's experiments, with an efficiency of $H = 0{\cdot}6$. Consequently, Table XII, the estimated molybdenum savings given and implied in his paper, covers idealistic conditions only. For incomplete hardening, Mr. Wilcock's figures leave no doubt that the boron steels are much inferior to the boron-free in regard to toughness. This, to my mind, is the most disturbing feature of the whole picture.

One cannot against this background dismiss the warning given by Mr. Wilcock that " in large sizes in which neither the boron-free nor the boron-treated steels harden fully, tensile properties are similar, but the impact values obtainable on the boron steels are inferior—usually very much so—to those on the boron-free steels." As he says, " The use of boron in alloy conservation would therefore demand a more delicate splitting up of specifications into size groups than is usual at present." For general-purpose steel usage, it seems that any such revised specifications should also have some stipulation covering the efficiency of quench.

Finally, a brief word about the molybdenum production situation, to clear up the prevalent misconceptions concerning molybdenum resources. During 1952 the production at Climax, Colorado, of nearly 24,000,000 lb. of molybdenum contained in concentrate was almost 100% greater than in 1950. Even more important, the end of 1953 is expected to see the culmination of a huge

Table A

TENSILE PROPERTIES OF BORON-FREE AND BORON-TREATED STEELS

(Heat-treated in the form of $\frac{5}{8}$-in. dia. bar)

Steel	Heat-Treatment	U.T.S., tons/sq. in.	Proof Stress, tons/sq. in.			Elongation, on $4\sqrt{A}$, %	Reduction in Area, %	Izod Impact, Value, ft.lb.
			0·2%	0·1%	0·05%			
	Ultimate Tensile Stress Approximately 65 tons/sq. in.							
S.A.E. 4140	Furnace-cooled 860° C.	50·0		24·8*		23·4	38·8	...
	Oil-quenched 860° C., tempered 640° C.	67·3		59·5*		22·0	54·9	49, 48, 48 } average = 48
S.A.E. 1035 + boron	Furnace-cooled 860° C.	37·5		19·2*		35·3	53·9	...
	Oil-quenched 860° C., tempered 500° C.	66·6	60·9	60·4	60·2	19·5	56·7	42, 41, 43 } average = 42
S.A.E. 1035	Furnace-cooled 860° C.	37·2		19·9*		34·0	45·9	...
	Water-quenched 860° C., tempered 470° C.	67·7		61·0*		18·0	55·7	47, 48, 50 } average = 48
	Ultimate Tensile Stress Approximately 90 tons/sq. in.							
S.A.E. 4140	Oil-quenched 860° C., tempered 470° C.	93·2	83·3	81·9	80·7	12·0	35·5	10, 8, 10 } average = 9
S.A.E. 1035 + boron	Oil-quenched 860° C., tempered 390° C.	87·5	80·8	80·3	79·0	16·0	51·8	10, 13, 14 } average = 12
S.A.E. 1035	Water-quenched 860° C., tempered 320° C.	95·5	83·7	82·7	81·2	10·0	30·0	4, 4, 4 } average = 4

* Average values

Table B
ANALYSES OF STEELS TESTED

Steel	C, %	Mn, %	Si, %	Cr, %	Mo, %	B (sol.), %
S.A.E. 4140	0·42	0·96	0·27	0·98	0·22	...
S.A.E. 1035 + boron	0·36	0·86	0·23	...	...	0·0023
S.A.E. 1035	0·35	0·86	0·27	...	...	...

expansion scheme there, on which $35 million has been spent; indeed, the major part of these new facilities is now in operation. Output this year is expected to exceed 36,000,000 lb. of molybdenum. Although supplies still remain under full allocation, and there is no excuse for extravagant usage, the general production position for the future is far better than it has ever been.

Mr. W. E. Bardgett (The United Steel Companies, Ltd.): Mr. Knowlton's statement that fatigue test results obtained on finished parts were essential information is, I think, most important. I have always felt doubtful about the value of the extensive work that is normally carried out on laboratory fatigue specimens. We can determine fatigue properties of the material, the effect of various surface conditions on the fatigue strength, and the effects of stress concentration, but any measurement we make concerning the latter is generally fairly arbitrary, being unrelated to a particular design. I believe that in future much more work will be done on completed components, because so much depends on design and on the stress concentrations involved.

In considering the fatigue of the substitute boron-treated spring steels, is there any possibility that the substitute steels are less liable to decarburization than the silico-manganese steel ? With regard to impact and its relation to fatigue, we have carried out tests on temper-brittle material treated to give the two conditions, tough and brittle, with impacts of about 60 ft.lb. and 8 ft.lb. respectively, and in both conditions this material showed precisely the same fatigue strength, both unnotched and notched.

I assume that the lower tempering temperature of the substitute steels may be one reason why special care is necessary in heat-treatment. There is also the point that if quenching is not entirely satisfactory there may be loss in properties, as has been indicated by Mr. Wilcock. The particular care that is being exercised in heat-treatment will act in favour of the boron steels, and it would be interesting to know whether the author considers this to be a factor of some importance.

I should be grateful if Mr. Knowlton would explain the significance of the depth hardenability test on carburized specimens. After the specimen is carburized and end-quenched, its hardenability is determined in the normal manner along the length of the specimen at various depths below the surface, and this shows the increased hardenability of the core and the lower hardenability of the higher-carbon case. In quenching a component, however, heat is extracted in a direction at right-angles to the case, not from one end as in the Jominy specimen, and it is not clear, therefore, how the results obtained at various levels below the surface are interpreted in terms of practical heat-treatment.

Mr. Wilcock's paper is an excellent assessment of the effect of boron in the manganese–molybdenum type of steel. It shows that the addition of boron results in an increase of ruling section for specific mechanical properties, but that beyond a certain size the properties are inferior to those of the boron-free steel. According to Mr. Wilcock, the advantage of adding boron is relatively small but quite definite. This indicates that, although boron permits upgrading in a narrow range, it could not be applied fully to upgrade a steel on the basis of our specifications and would mean, as Mr. Wilcock said, splitting up into size groups. In what way can we take advantage of the increased hardenability arising from boron addition ? Mr. Brinley Jones indicated that if we apply it to a particular ruling section we can design a composition very definitely to effect economy in alloys, but in such a case it would be necessary to consider only requirements where full casts were involved.

We have been working on the application of boron to a steel for cold-headed bolts, and I think Table *A* may be of interest. In this Table, a Cr–Mo bolt steel, oil-quenched and tempered at 640° C., is compared with S.A.E.1035 boron-treated steel, oil-quenched and tempered at 500° C. to a tensile strength of about 65 tons/sq. in. The compositions of the steels are given in Table *B*. There is no significant difference between the properties of the two steels heat-treated in the form of $\frac{5}{8}$-in. dia. bar. The results show further that an S.A.E. 1035 boron-free steel requires to be water-quenched to give equivalent properties.

Table *A* shows also the same steels hardened and tempered to a tensile strength of about 90 tons/sq. in. To give this tensile strength, the Cr–Mo steel was tempered at 470° C. and the boron steel at 390° C. There is very little difference between the properties of the alloy steel and the boron steel, although there is a slight advantage in favour of the latter in respect of elongation, reduction in area, and impact, in spite of the lower tempering temperature adopted for this steel. The boron-free carbon steel, which was water-quenched and tempered at a temperature of 320° C., differed particularly from the other two steels in having a much inferior impact value.

Figure *A*(i) shows the hardenability of the alloy and boron-treated steels. Although the hardenability of the

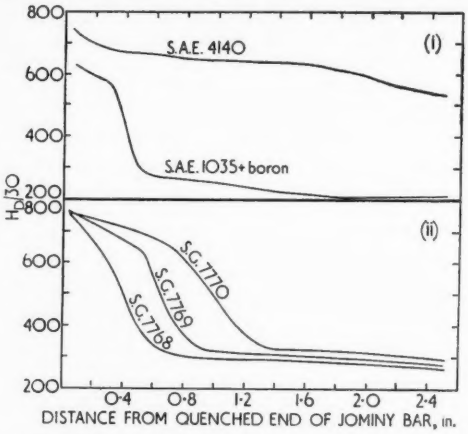


Fig. A—Hardenability curves for (i) alloy and boron-treated steels (sub-standard $\frac{1}{2}$-in. dia. specimens), (ii) boron-free and boron-treated steels

Table C

TENSILE PROPERTIES OF A 0·45–0·55% C, 1·5–1·8% Mn STEEL (Heat-treated in the form of 1½-in. dia. bar)

Cast No.	C, %	Mn, %	B, % (sol.)	Method of Adding Boron	Heat-Treatment	U.T.S., tons/sq. in.	20% Proof Stress, tons/sq. in.	Elongation, on $4\sqrt{A}$, %	Reduction in Area, %	Izod Impact Value, ft.lb.
S.G.7768	0·45	1·44	...	...	Oil-quenched 860° C., tempered 550° C.	59·2	42·0	18·0	41·8	11
S.G.7769	0·47	1·48	0·0027	Silcaz	Oil-quenched 860° C., tempered 590° C.	58·7	49·4	20·0	52·0	37
S.G.7770	0·46	1·60	0·0029	Aluminium ferro-boron	Oil-quenched 860° C., tempered 590° C.	60·0	50·4	20·0	51·0	33

boron-treated steel is very much inferior to that of the alloy steel, it is adequate to ensure satisfactory mechanical properties in sizes up to ⅝-in. dia.

In considering this material for cold-headed bolts, quite apart from the question of alloy conservation there is the advantage, as shown by the results in Table *A*, that the boron steel can be softened to a much greater extent than the alloy steel and in consequence should lead to much improved die life.

Table *C* shows the properties of a 0·45–0·55% C, 1·5–1·8% Mn steel, which is required to give a proof stress of 45 tons/sq. in. in the oil-quenched and tempered condition and in the form of 1½-in. dia. bar. With the lower limits of carbon and manganese the required proof stress is not obtained, but when boron is added the proof stress increases to 49·4 tons/sq. in., which is well within the specification. In addition, there is improved ductility, the elongation increases, and the reduction in area rises from 42 to 52%, and also the impact value rises from 11 to 37. It is noteworthy that the boron-treated steel was tempered at a temperature of 590° C. compared with 550° C. in the case of the boron-free steel. The results for S.G. 7770, however, although not strictly comparable with those for the boron-free steel, owing to the somewhat higher manganese content, tend to confirm the beneficial effect of boron, particularly in regard to the higher impact value for equal tensile strength.

Figure *A*(ii) shows the hardenability of the boron-free steel, the Silcaz-treated steel, and the aluminium ferro-boron treated steel. These curves clearly illustrate the advantage to be gained by the boron addition. The difference in hardenability of the two boron-treated steels is no doubt largely attributed to the higher manganese content of the ferro-boron treated steel. This example shows clearly that when a steel is borderline when working to the lower end of the analysis range, the addition of boron would ensure satisfactory mechanical properties; moreover, to obtain such an improvement by change in composition would mean the use of a substantial amount of, say, molybdenum—of the order of 0·10%.

These two sets of results, which were obtained on material from an 18-lb. H.F. furnace, have been selected from a number of experiments on boron steels to illustrate applications which we have considered as meriting development, and, from experience gained in the laboratory and in production, we consider that laboratory test results can be satisfactorily reproduced in full-scale production.

Mr. **A. J. K. Honeyman**: Whilst I do not feel competent to discuss Mr. Knowlton's paper, I would express my appreciation of the manner in which he has described his experience of the use of boron steels in the U.S.A. It certainly appears that American steelmakers are convinced that boron is effective practically.

Our approach in this country is more cautious, and Mr. Wilcock is therefore to be congratulated on his very complete study of the effect of boron on a particular type of low-alloy steel. As he indicates in his introduction, hardenability is not a sufficient criterion of the effects of boron on mechanical properties. (At least, I thought so before hearing Mr. Knowlton.) All that hardenability determinations can do is to serve as a guide to the probable ruling section. Figure *B*, taken from a paper published in *Metal Treatment*,[2] illustrates this. Up to 0·5 in. dia. there is obviously no need to use boron; from 0·5 in. to 2·5 in. dia., since the boron-treated steel is fully hardened and the non-treated steel is not, boron would be expected to be beneficial in all respects. Between 2·5 in. and 4·0 in. dia., the greater hardenability of the boron-treated steel may or may not result in improved mechanical properties. It is in this range, therefore, that tests on the lines conducted by Mr. Wilcock should be carried out, with a view to the possibility of increasing the ruling section. Beyond 4 in. dia., boron is likely to be detrimental.

In the same paper,[2] I referred to the fact that the technique employed to detect boron microscopically consists essentially of isothermal transformation in the pearlitic range. I therefore suggested that the "mechanical properties are definitely improved provided advantage is taken of the increased hardenability obtained and the steel is fully hardened before tempering. With partial hardening, boron is beneficial provided no transformation takes place in the pearlitic range." Mr. Wilcock states that his "limited data, therefore, suggest that the use of boron might have the effect of raising the impact transition temperatures." From his own hardenability curves and hardness-traverse data it is evident that the boron-treated steels were not *fully* hardened, and I would suggest that the lower impact properties obtained by him are due to partial transformation in the pearlitic range.

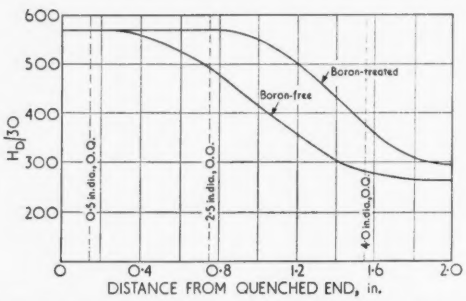


Fig. B—Hardenability curves for boron-free and boron-treated steels

Glen[8] showed that boron has a considerable effect on all steels except plain manganese steels containing less than 0·3% of carbon. With increasing total alloy content, *i.e.*, in the more complex steels, the effect of boron tends to become extremely large.

Since these complex alloy steels do not readily transform in the pearlitic range, it is the properties of such steels which should be investigated if a substantial conservation of alloying elements is to be achieved by the use of boron.

Dr. **N. P. Allen** (National Physical Laboratory): The statement that fully hardened and tempered steels, whatever their composition, have the same mechanical properties if they are tempered to the same value of tensile strength is broadly true in regard to the proof stress and elongation, but it is only roughly true of the reduction in area, and not even roughly true of the impact value. It is very important that this should be fully understood.

I have recently seen results on a number of 1½% Mn–Mo steels of nearly identical composition, all of which had been heat-treated in bars of the same size and fully hardened and tempered to equivalent tensile strengths. According to the above statement, all the steels should have had the same impact value, but in fact the values at 0° C. ranged from about 80 ft. lb. to about 20 ft. lb., and at — 78° C. from about 50 ft. lb. to 7 ft. lb. This difference was entirely due to the absence of phosphorus in those steels which gave the high values; the other steels contained a normal amount of phosphorus (about 0·05%).

Boron is said to have a similar effect, *i.e.*, it lowers the impact values, under certain conditions. An important question is therefore whether high impact values are necessary. This has yet to be decided, and should be determined as soon as possible.

I think Mr. Knowlton will agree that if the steel is likely to undergo plastic deformation, high fracture stress is necessary. The application of steels must be divided into those where plastic deformation is normally expected and those where it is not expected to occur in a perfectly designed component, and it is the latter which involve most difficulty. Most of the engineering components used in the automotive industry are designed to be well within their elastic limit all the time, and they fail, if they fail at all, usually by fatigue. Unfortunately, we do not know whether there is any relation between the weakness exhibited by the impact test and the behaviour in fatigue. Scientific study of fatigue indicates that plastic deformation must take place before the material can suffer from fatigue, and yet the greatest difficulty with notch fatigue is found in those steels which have a very high ratio of elastic limit to tensile strength, and where the fatigue limit is usually well below the elastic limit as normally developed, *i.e.*, where the material is not expected to undergo plastic deformation.

There is a contradiction here, however, in that when a component has failed by fatigue it has surely undergone plastic deformation somewhere. If a steel is in a condition such that when it undergoes plastic deformation it is also near the limit of its brittle strength, it is rather dangerous, but there is not yet any experimental evidence of a relationship between the Izod impact value as normally determined and the behaviour of any material in fatigue.

Mr. **G. T. Harris** (Wm. Jessop and Sons, Ltd.): Mr. Knowlton has drawn attention to the fact that boron affects mechanical properties other than hardenability, but so much has been said about the latter that it may be as well to consider the other important effects. I wish to refer to the highly alloyed austenitic steels used in gas turbines, containing up to 50% of alloying element. We have carried out experiments on a steel used for jet-engine discs and having the following composition: C, 0·4%; Ni, 13%; Cr, 13%; Co, 10%; W, 2·5%; Mo, 1·8%; Nb, 3·0%. Figure *C* shows the effect of boron additions on the 100-hr. hot rupture strength of this steel at 700° C. Addition of only 0·01% of boron has raised the strength by 25%, an increase which could have been attained otherwise only by adding very considerable amounts of expensive alloying elements.

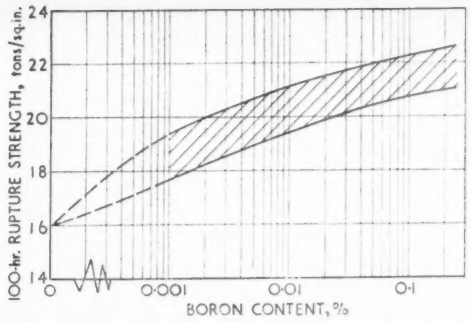


Fig. C—Effect of boron addition on 'creep strength'

I should like to ask Mr. Knowlton whether he knows of any American work which throws light on how this increase of hot strength is effected. We have noticed, for instance, in following by change of electrical conductivity the ageing of such steels, that the rate of ageing is altered by boron. The time for full ageing at 750° C. was 60 hr. for the boron-free steel and 100 hr. for a steel with 0·03% of boron. Correspondingly, the ageing rate when the material was half aged was lowered from 0·050%/hr. to 0·030%/hr.

Dr. **C. L. M. Cottrell** (British Welding Research Association): One of the special features of low-carbon boron–molybdenum steels is good weldability combined with a high yield strength. This type of steel has a high resistance to heat-affected zone cracking, even when welded with electrodes having a high effective hydrogen content in their coatings. I should like to say a few words about what I feel to be the reason for these good properties.

We at B.W.R.A. have recently examined the transformation characteristics during continuous cooling of 34 experimental low-alloy Ni–Cr–Mo steels and of a commercial B–Mo steel, using a rapid-action dilatometer. In this dilatometer the thermal cycle occurring in the heat-affected zone of Reeve fillet-weld cracking tests on the steels was simulated. High-frequency induction heating and gas-quenching were used to produce the rapid thermal cycle in the dilatation specimen.

It was found that there was a critical end-of-transformation temperature, *i.e.*, temperature at which the steel finished transforming during continuous cooling, neglecting any retention of austenite down to room temperature. If a steel continued transforming below this critical temperature in the rapid dilatation test, heat-affected zone cracks were observed in the corresponding weld-cracking test. The critical temperature was 290° C. for welds made with rutile-coated electrodes, *i.e.*, B.S.1719 class E217.

The relationship between cracking in the weld test and end-of-transformation temperature in the rapid dilatation test for the 35 steels is shown in Fig. *D*. The B–Mo steel *X* has a high end-of-transformation temperature, about 425–430° C., compared with those of the Ni–Cr–Mo steels of similar mechanical strength, *e.g.*, 255° C. for steel 14; also, steel 14 cracks and steel *X*

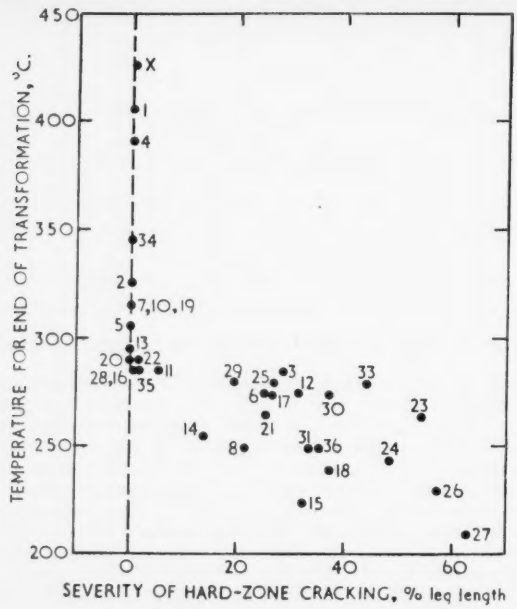


Fig. D—Relationship between end-of-transformation temperature during continuous cooling and the extent of cracking in the weld test, for 35 steels

does not. The full results of this work have recently been published.[4]

We are now investigating the relationship between end-of-transformation temperature and cooling time (or cooling rate) for several commercial steels. As would be expected, there is a marked lowering of the end-of-transformation temperature as cooling rate is increased. This effect varies as the composition of the steel is altered.

Figure *E* shows results for two steels of similar strengths whose plate analyses were as follows:

	C, %	Mn, %	Mo, %	B, % sol.	B, % insol.
Mn–Mo steel	0·19	1·3	0·25	...	...
B–Mo steel	0·10	0·55	0·48	0·0022	0·0018

Cooling times between 4 and 100 sec. cover the range found in normal metal-arc welding. Despite the similarity in strength, the end-of-transformation/cooling-time curves for the two steels are quite different. I believe this difference explains the superior weldability of the B–Mo steel, which is weldable with complete freedom from heat-affected zone cracking (using a rutile-coated electrode) at all cooling rates shown, whereas the Mn–Mo steel has been shown to crack with cooling rates greater than 6° C./sec. at 300° C. (*i.e.*, cooling times less than 30 sec.).

It should, however, be mentioned that when low-hydrogen electrodes are used for welding, the critical temperature can be lowered considerably, to below 250° C., so that good weldability is obtained with the Mn–Mo steel. At the same time, the B–Mo steel is weldable with a general-purpose electrode. This result agrees with those of Fig. *D*.

The difference between the curves for the two steels is probably due mainly to the difference between their carbon contents. Therefore, the good weldability of the B–Mo steel can be attributed largely to its low carbon content. The addition of boron gives a much increased rate of transformation in the intermediate range, and this effect presumably contributes to the strength of the steel by promoting an increased amount of this type of transformation at the cooling rates prevalent during manufacture. At the same time, the intermediate transformation is so rapid that, even at the high cooling rates associated with welding, transformation is completed above 300° C. In other words, the effect of boron in this steel is most marked at lower cooling rates, which is very desirable from the viewpoint of weldability. I should like to ask the authors why boron has this effect on the intermediate transformation.

Mr. **L. H. Toft** (Mond Nickel Co., Ltd.): I have recently studied the effects of boron treatment on the impact properties of N.E. 8630 and N.E. 8640 steels in both the fully hardened and partially hardened conditions. Boron-free and boron-treated samples of these steels, containing approx. 0·85% Mn, 0·45% Ni, 0·5% Cr, and 0·2% Mo, had been obtained by my Company from the U.S.A. during the war. Jominy end-quench tests confirmed that boron treatments had increased the hardenabilities of these two types of steel. Izod test specimen blanks, 0·47 in. dia., were machined from $\frac{5}{8}$-in. dia. bars of each steel and were fully hardened by oil-quenching from 850° C. Further blanks were cooled in still air after soaking at 850° C., to simulate the cooling of the axes of 4-in. dia. bars quenched in oil. This treatment only partially hardened each of the steels. The fully hardened and partially hardened specimens were tempered for 1 hr. at temperatures between 200° and 700° C. and water-quenched. Standard 0·45-in. dia. Izod impact bars were ground from the heat-treated blanks, notched, and broken at room temperature. The test results are given in Figs. *F* and *G*.

When tempered at temperatures below 400° C., *i.e.*, to hardness levels above about 410 D.P.N., the boron-treated N.E. 8630 steel in the fully hardened condition has better impact properties than the boron-free steel (Fig. *F*(i)). The improvement is greatest at a tempering temperature of 300° C., the impact value being raised from 14½ to 25 ft. lb. On the other hand, when tempered to hardness levels below about 400 D.P.N. the impact values of the boron-treated steel are, on the average, about 8 ft. lb. lower than those of the boron-free steel. With the higher-carbon N.E. 8640 steel in the fully hardened condition (Fig. *F*(ii)), the effect of boron at high hardness levels is not so marked, but the greatest improvement again occurs at a tempering temperature of 300° C., the impact value being raised from 9½ to 14

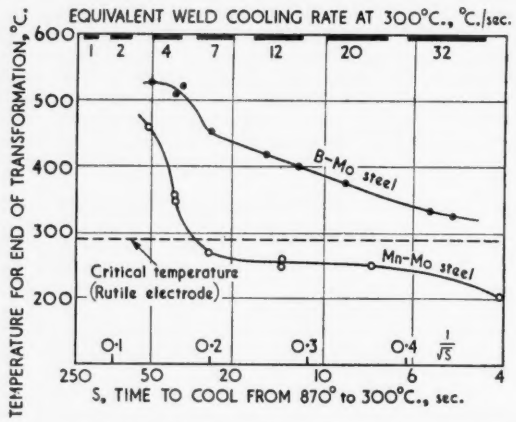

EQUIVALENT WELD COOLING RATE AT 300°C., °C./sec.
TEMPERATURE FOR END OF TRANSFORMATION, °C.
S, TIME TO COOL FROM 870° to 300°C., sec.
B-Mo steel
Mn-Mo steel
Critical temperature
(Rutile electrode)

Fig. E—Relationship between end-of-transformation temperature and cooling time for two steels of similar mechanical strength (0·3% proof stress = 30 tons/sq. in. in normalized ½-ton thick plate)

ft. lb. After tempering at 400° C. and above, the impact values of the boron-treated steel are again about 8 ft. lb. lower than those of the boron-free steel.

This effect has been reported[5, 6, 7] many times. The above results agree with those presented by Mr. Wilcock in showing the greatest improvement in impact properties after tempering at 300° C. Mr. Wilcock showed that the advantage gained by the boron treatment of N.E. 8640 steel is lost after tempering at 450° C. and above. The results I have given here go further and suggest that in this tempering range the boron-treated steels have lower impact values than the corresponding boron-free steels. The reduction is significant but cannot, of course, be regarded as serious, since all the impact values are in excess of 30 ft. lb.

Most of the partially hardened and tempered Izod bars show large variations in impact values from notch to notch on the same bar, a variation of 20 ft. lb. frequently being observed. These variations may be due to differences in the orientation of the tempered martensite areas in the microstructure relative to the notch.

Figure *G* shows the effect of boron on the impact properties of N.E. 8630 and N.E. 8640 steels when partially hardened and tempered. The vertical lines represent the range of impact values obtained from each specimen. Although the boron-treated N.E. 8630 steel (Fig. *G*(i)) hardens to a slightly greater extent than the boron-free steel (its average hardness before tempering being 285 D.P.N. whilst that of the boron-free steel is 257 D.P.N.), when tempered to the same hardness levels its impact values are appreciably lower than those of the boron-free steel.

The hardnesses of the boron-treated and boron-free N.E. 8640 steels (Fig. *G*(ii)) before tempering were 400 D.P.N. and 301 D.P.N., respectively. The scatter of results is such that average lines through the plotted values cannot be justified. It is clear, however, that no significant difference is revealed between the impact values of the boron-free steel and those of the boron-treated steel, despite the greater hardness of the latter steel before tempering.

These results again only partially confirm the conclusions reached by Mr. Wilcock. Although he states that, where neither the boron-free nor the boron-treated steels harden fully, impact properties obtained on the boron steels are inferior to those obtained on the boron-free steels, it is obvious that in the above instance, where there is a difference of 100 D.P.N. in the as-quenched hardnesses of the boron-free and boron-treated steels, the detrimental effect of boron on the impact values in the partially hardened and tempered condition has been offset by its effect of increasing the as-quenched hardness.

A study of the fracture appearances of the broken Izod bars resulting from the tests just described leads me to assume that, in the fully hardened and tempered condition, boron lowers the impact transition temperature

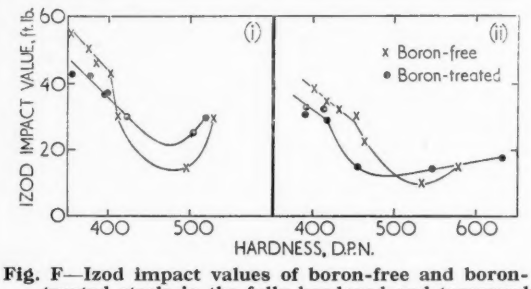


Fig. F—Izod impact values of boron-free and boron-treated steels in the fully hardened and tempered condition: (i) Steel N.E. 8630; (ii) steel N.E. 8640

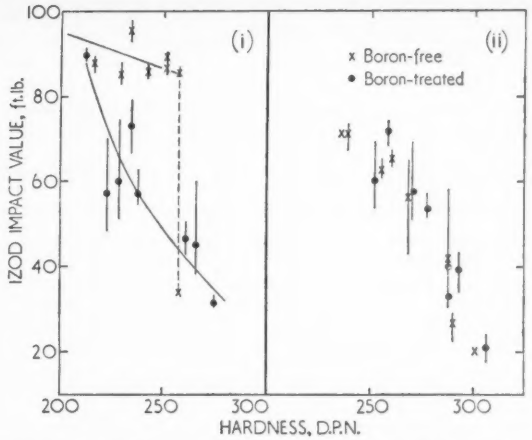


Fig. G—Izod impact values of boron-free and boron-treated steels in the partially hardened and tempered condition: (i) Steel N.E. 8630; (ii) steel N.E. 8640

of a steel tempered below 400° C. but raises the transition temperature when it is tempered above 400° C., and that in the partially hardened condition it usually raises the transition temperature.

Mr. **P. O. Bjorkman** (AB B. A. Hjorth and Co.): One subject that has not been discussed here is the machinability of boron steel. If, for example, drop forgings weighing up to 15 lb. and with dimensions up to 1 in. are to be made of alloy steel and then machined, they must be normalized and annealed to obtain the best machinability. Perhaps replacing these steels by ordinary C–Mn–B steels will improve the machinability.

The Jominy test covers all rates of cooling. This discussion has dealt mainly with rapid cooling, but with slow cooling perhaps the boron steels give lower hardness, and therefore better machinability, than alloy steels with the same hardenability.

Mr. **R. J. Brown** (Morris Motors, Ltd.): The question of distortion is a very serious factor in the passenger-car industry. In the heavier parts which Mr. Knowlton produces for trucks and tractors, is the noise in the final assembly such that it can be tolerated, whereas a similar noise level in a passenger vehicle would not be tolerated, and is this the reason why the boron-treated steels are used in the heavy-vehicle industry but have found little, if any, use in the passenger-vehicle industry in the U.S.A. ?

It has been stated that it is necessary to have a closer degree of control over the heat-treatment, and reference has been made to the effect of slack-quenching and to a 25% drop in the notch fatigue strength. Are we to assume from this that slack-quenching does occur in practice in the U.S.A. ? My opinion is that slack-quenching is not usually tolerated in this country, because of its effect on notched-bar impact results.

Are the boron-treated steels of finer grain size than the equivalent alloy steels, and is there any danger of machining difficulties arising from the adverse effects of the extra additions of aluminium ?

Mr. **D. Hardwick** (B.S.A. Group Research Centre): We have carried out a few preliminary tests on two basically similar steels, one with and one without boron, with the compositions given in Table *D*. Each steel was melted in a 500-lb. induction furnace and deoxidized with silico-manganese and aluminium. To the boron heat sufficient ferro-boron was added to give a boron addition

Table D
COMPOSITIONS OF EXPERIMENTAL STEELS

Steel	C, %	Mn, %	Si, %	S, %	P, %	Ni, %	Cr, %	Mo, %	B, %
Boron	0·44	0·71	0·34	0·032	0·010	0·29	0·59	0·05	0·0028
Boron-free	0·46	0·72	0·31	0·029	0·010	0·34	0·75	0·05	...

of 0·005% and a boron content in the steel of 0·003%. Slight variations in composition, particularly chromium, occurred, owing to the impossibility of making a split heat.

Figure *H* shows the results of Jominy end-quench hardenability tests on the two steels in the as-quenched, tempered 300° C. and tempered 650° C. conditions. These curves agree well with those obtained by Mr. Wilcock. At distances up to 0·2–0·3 in. from the quenched end of the Jominy bar, the boron steel is the softer, but then a definite increase in hardenability is apparent. At normalizing rates of cooling and slower, the boron steel is again the softer (*see* Fig. *H*), which is not surprising in the absence of molybdenum. The difference in hardenability as reflected in the hardness results is maintained after tempering at 300° C., but disappears at 650° C. Microscopical examination of the test bars revealed the presence of 100% martensite up to 0·25 in. and 0·45 in. from the quenched end in the non-boron and boron steel, respectively. Assuming a severity of quench number $H = 0·35$/in., bars of 1¼ in. dia. of the boron steel and ¾ in. dia. of the non-boron steel would through-harden when oil-quenched.

Mechanical test results on the two steels heat-treated as 1¼-in. dia. bar are given in Fig. *I*. It will be noticed that in this size the boron steel would through-harden and the non-boron steel would not. At tempering temperatures of 400° C. and below, the boron steel has a higher tensile and proof stress; above this temperature the curves are identical. The impact values show the superiority of the boron steel at all tempering temperatures. The results of elongation and reduction in area tend to be erratic, but in all cases show the advantage of the boron steel.

Dr. **L. F. C. Northcott** (Armament Research Establishment, Woolwich): The information on the effect of boron additions on grain size given in Mr. Rohl's paper this morning (*see* p. 179) suggests that the boron steels had larger grain size than the boron-free. I should like to know whether the effect of the increased addition of aluminium is intentional, to counteract the grain-coarsening effect of the boron, or whether the greater hardenability is due to the coarser grain of the boron steels.

In a 150- or 100-ton cast, where up to 40 ingots have been prepared, what are the mechanical properties (under identical heat-treatment conditions) of material from various parts of the first, middle, and end ingots, without further aluminium addition? Is it necessary to add more boron or aluminium at various stages during the casting operation?

The trials carried out in this country and which have been described have been on H.F. melts varying from 50 lb. up to 6–10 cwt. It would be interesting to know what results would be obtained on a large cast of basic O.H. steel, of which 85% of the steel produced in this country is made.

I would like to refer to one Figure of Mr. Knowlton's paper in which the hardenability band represented an average of over 100 casts of nominally the same steel. This type of spread is not very good, and for military applications would be wholly unacceptable.

I hope that manufacturers in this country will place less emphasis on the importance of boron from the standpoint of hardenability. Engineers and Service departments do not base their work on hardenability but on proof stress; they prefer good elongation or ductility and high notch-bar impact values. Proof stress is essential, but hardenability is not a quality on which the Service departments could rely.

Mr. **M. A. Kohn** (I.R.S.I.D.): We have recently studied five industrial casts of boron steels of the following compositions:

	Equivalent A.I.S.I. Grade	C, %	Mn, %	Ni, %	Cr, %	Mo, %
A	94B17	0·17	0·80	0·65	0·58	0·18
B	80B15	0·19	0·79	0·24	0·30	0·24
C	81B35	0·34	0·85	0·32	0·44	0·17
D	81B35	0·36	0·72	0·27	0·50	0·12
E	86B45	0·44	0·98	0·54	0·60	0·12

Boron: 0·0015–0·0020%.

The tests showed that the carburizing grades (steels *A* and *B*) have good mechanical properties after oil-quenching and tempering at low temperatures, and that they may be carburized by pack- or gas-carburizing. In

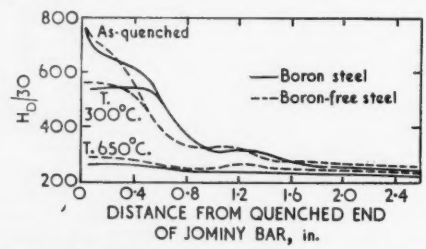


Fig. H—Hardenability curves for boron-free and boron-treated steels in the as-quenched, tempered 300° C. and tempered 650° C. conditions

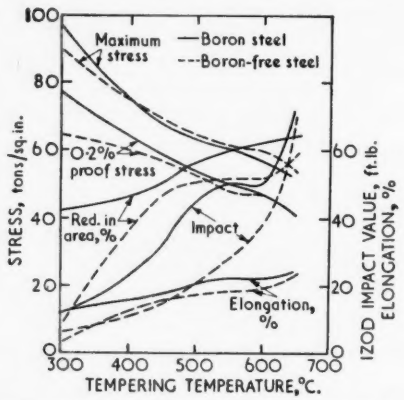


Fig. I—Tempering curves for boron-free and boron-treated steels, oil-quenched 860° C. (heat-treated as 1¼-in. dia. bar)

this respect they seem to be as good as the French conventional 1% Cr, 0·3% Mo steels, but they cannot be carbo-nitrided at the temperatures usual in automotive production. In addition, they need austenitizing temperatures higher than those of the conventional steels that they replace.

Some of our results for the medium-carbon alloy steels are in very close agreement with Mr. Wilcock's conclusions. Steels *C* and *E* were made in basic electric-arc furnaces. One-half of the molten steel was tapped into a ladle and boron was added as Grainal; the second half of the cast was tapped into another ladle without boron addition. It was thus possible to determine the exact effect of boron on casts made in accordance with industrial practice.

In small sections, boron has no effect on the mechanical properties of both these steels, for even the boron-free steel is given a martensitic structure by quenching, and then is fully hardened. After tempering up to 500° C., the mechanical properties at the centre of 15–20-mm. dia. rounds of the boron-treated and boron-free steels are similar, except for the yield-stress values, which are slightly higher in the boron-treated steel.

At the centre of 40-mm. dia. round bars, quenched and tempered at 500° C., the boron-treated steel *C* has better tensile strength, yield ratio, and impact-toughness than the boron-free steel, the elongation being similar. At the centre of 50–100 mm. (2–4 in.) dia. round bars, quenched and tempered at 550° C., the boron-treated steel *E* has better tensile strength, yield ratio, and impact toughness than the boron-free steel quenched and tempered at 500° C. The percentage elongation and the reduction in area of both steels are very similar.

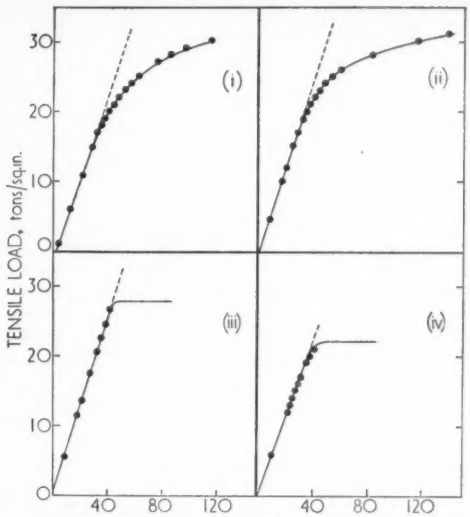
In large sections (8-in. dia. rounds) boron-treated steel has no significantly better properties than boron-free steel except near the surface, where the cooling rate is high enough to allow the boron to have a favourable effect. After tempering at 650° C., boron-treated steel has slightly better tensile strength and yield-point than boron-free steel, although the difference is not very noticeable.

As in our previous investigations on industrial casts, these results show that boron has a favourable effect in every case when boron-treated steel is given a martensitic structure which is not produced in the boron-free steel. Thus, by boron addition, it is possible to increase the properties of the steels in intermediate sizes, the limits of which vary according to the alloying-element contents.

Mechanical properties, however, are not the only point to be considered. In France it was formerly thought, from American information, that the effect of boron was the most favourable in carburizing grades for parts used in the automotive industry. However, tests made with the co-operation of two important automotive firms showed that these steels were not convenient for such parts; they cannot be austenitized or carbo-nitrided at the temperatures generally used in continuous production. Moreover, the steels used in this industry already have very low nickel and chromium contents, and it does not seem possible to reduce them further.

It seems now that the most convenient boron grades are those used for heat-treated parts where the main characteristics wanted are a high tensile strength throughout the section, associated with good impact properties. Further tests are to be made in this connection.

Finally, the use of boron steels cannot be considered in the same way in France as in the United States, and the boron steels must be related to the particular conditions existing in each country. Moreover, boron steels seem to have properties that may be useful for some special applications apart from the purpose of saving alloying elements, and tests are to be made to investigate all their possible uses.



	C, %	Ni, %	Mo, %	Boron
(i)	0·19	2·0	0·23	...
(ii)	0·20	2·0	0·09	Yes
(iii)	0·21	2·0	...	Yes
(iv)	0·18	2·0	...	...

Fig. J—Tensile stress/strain curves for normalized boron-free and boron-treated steels

Mr. **G. Mayer** (Mond Nickel Co., Ltd.): During the course of some work several years ago on the effects of boron in nickel-containing steels, an interesting effect was observed which did not then appear to merit further attention, but might now indicate a very useful application of boron. The effect was that of raising the elastic limit of a 0·2% C steel of the weldable type. At the time of the investigation, emphasis was placed on the yield strength of high-tensile welding steels, and a 0·2% C, 2·0% Ni, 0·2% Mo steel was one of the best available. This steel had a 0·5% proof stress value of about 30 tons/sq. in. in the as-rolled condition or when normalized as ½-in. thick plate, but the yield was of the continuous type and the elastic limit was low. The latter could be raised, however, by tempering at 450° C. The stress/strain curve for this steel is shown in Fig. *J*, curve i.

Tests on experimental melts showed that one-half of the molybdenum in this steel could be replaced by boron without loss of mechanical or welding properties, the boron being added as Silcaz, preceded by deoxidation with 0·1% of aluminium. The stress/strain curve (Fig. *J*, curve ii) for the boron-treated steel is almost identical with that for the boron-free steel with higher molybdenum content. When all the molybdenum in the steel was replaced by boron, some loss in the yield strength was noted (*see* curve iii), but the yield was discontinuous and the elastic limit was high, about 28 tons/sq. in. This is considerably higher than the elastic limit of curves i and ii. In comparison, curve iv for a boron-free, 2% Ni steel is somewhat similar in that it has a sharp yield-point, but the elastic limit is lower.

The value of boron in raising the yield strength of weldable alloy steels has been demonstrated by Mr. Bardgett and Dr. Reeve, but I feel that greater attention should be given to the elastic limit of such steels, and this may well lead to further development.

In discussing the effect of boron on the mechanical properties of steels where its effect on hardenability is not involved, as when considering fully hardened and tempered steels, the type of boron-addition alloy used should not be overlooked. For example, it has been claimed that boron improves the impact resistance of fully hardened and lightly tempered steel. On testing this claim we found that a small increase in impact value was effected when the proprietary alloy Silcaz was used, but a similar increase was also obtained on using a synthetic Silcaz mixture, boron-free. Hence this effect may not be due to the boron but may be related to deoxidation practice, since a more intense deoxidation is usually carried out when boron is being added.

So far there has been only slight mention of the effect of boron on the mechanism of transformation. In this connection we have studied three laboratory-made steels containing 0·3% C, 0·7% Ni, 0·6% Cr, and 0·25% Mo. All three steels were treated with 0·1% of aluminium. In one case this was followed by an addition of 0·003% of boron, added as Silcaz; in the second melt the same amount of boron was added as manganese–boron; and the third steel was boron-free. Figure *K* shows the progress of transformation when these steels are isothermally transformed at 600° C. (The times shown refer to the holding times at that temperature.) In these micrographs, crystals of the precipitating pro-pearlitic ferrite are shown in a practically unetched martensite matrix. The effect of boron is clearly shown as markedly reducing the rate of nucleation of the ferrite crystals, but the rate of growth of each crystal is little affected. Figure *K* also shows that when an addition of titanium is omitted the effect of the manganese–boron addition is much less than that of Silcaz. This is obviously due to those elements in Silcaz that remove nitrogen, of which titanium is one.

The retardation of nucleation by boron was also observed at other transformation temperatures, and Fig. *L* shows the progress of transformation at 500° C.

In regard to the use of boron steels in the U.S.A., I believe that some figures published a few years ago gave the estimated annual production of boron steels in 1952 as between 5 and 10 million tons, but only about 700,000 tons of boron-treated steel were made. Is this because most American steel users are not prepared to accept boron steels as completely satisfactory substitutes for the higher-alloy steels? If so, perhaps Mr. Knowlton will tell us what their objections are.

Mr. **T. R. Middleton**: English Steel Corporation, Ltd.): Production of boron steels in the U.S.A., at 8% of the total output of alloy steels for general engineering purposes, would involve a considerable tonnage and a not insignificant conservation measure. However, I had formed the impression, from published literature, that boron was used more extensively, and it therefore seems clear that not all boron steels covered by the tentative standard specifications have proved acceptable; it may be that difficulties not touched upon by Mr. Knowlton and Mr. Rohl have been encountered.

Dr. Northcott expressed surprise that no effort has been made to use boron in the basic steel industry in this country, which produces 85% of our total output. I suggest that, with the bulk of basic steels, the concern is more to reduce than to increase hardenability, and one would not expect this industry, with its comparatively small alloy-steel production, to make appreciable use of boron.

Referring to criticism that this country has failed to make use of boron as an alloy saver, it should be realized that the position here is very different from that in the U.S.A. Conservation, even to the extent so far achieved in the U.S.A., would not have been possible in the absence of much improved heat-treatment practice, involving the installation of special plant, which could be justified by the huge tonnages of steel absorbed for a particular component; we are concerned with a large number of comparatively small users, and, generally speaking, similar measures are not economically possible. Moreover, our existing specifications impose impact requirements but, as re-emphasized by Mr. Wilcock although boron may be beneficial in a certain size range, it may be detrimental in larger ruling sections; this introduces complications when, as is so often the case, end uses are not known when casts of steel are made. Under existing circumstances boron is often better avoided.

CORRESPONDENCE

Mr. **H. H. Jackson** (Standard Motor Co., Ltd.) wrote: From American literature and this discussion it is obvious that there is a field of application for boron-treated steels. We intend to explore this field with the co-operation of our steelmakers. Both they and the users can now benefit from the development which has occurred in America, where the difficulties and limitations of boron steels are more clearly recognized.

The difficulty in translating American practice to British practice is that boron improves hardenability only, and this is the basis of many American steel specifications. Existing British specifications happen to be of the compositions not greatly affected by boron, and the field of usefulness appears to be those instances where a slack-hardened steel can be fully hardened by boron addition. These requirements are not widespread in this country.

Whereas British specifications are based on the strength and Izod properties from the centre of a section, American specifications are based on the hardness of the centre, and the omission of Izod requirements is significant. In this country the trend is for hardenability to become more recognized as a specification requirement, although there has been no definite move for this property to replace any of our present physical test requirements. In any case, the trend is towards the lower-alloy steels which, incidentally, are most sensitive to boron treatment; and, since the low-alloy steels themselves will be more economic, their use will undoubtedly expand rapidly, during which time, I suggest, experiments with boron treatment could proceed with advantage. It appears, therefore, that low-alloy steels must first be made and tested, and then some of them must be treated with boron. This is being carried out to a certain extent in experiments with shallow-hardening or shell-hardening steels, and it seems that most useful practical work can be done on components such as bolts, gears, and axle shafts.

AUTHORS' REPLIES

Mr. **H. B. Knowlton** wrote in reply: The main objective of my paper is to show to what extent the industries in the U.S.A. have been able to effect a saving in critical alloys, particularly nickel and molybdenum, by the use of low-alloy boron-containing steels in place of conventional automotive steels, such as 8600, 4100, etc. The 80*B*00 Cr–Ni–Mo–B and the 50*B*00 Cr–B steels were created to duplicate the hardenability of such steels as 8600. Past experience, particularly during World War II, had shown that a new alloy steel selected on the basis of duplicating the hardenability of a standard steel would have the same response to the standard heat-treating process and would produce similar hardness values over a given cross-section, and consequently similar mechanical properties. Some exceptions to this statement are discussed in the paper; in general, however,

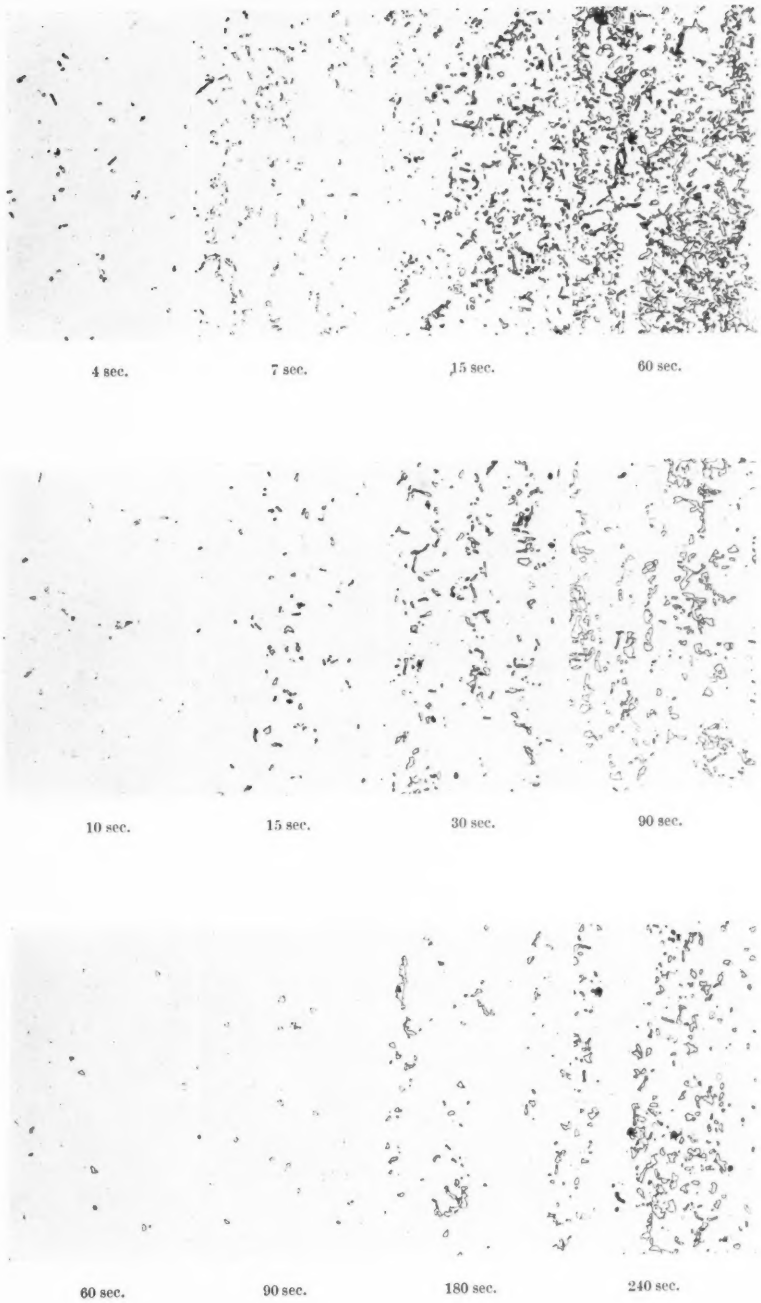


Fig. K—Microstructures of boron-free and boron-treated low-alloy steel partially isothermally transformed at 600° C. × 200

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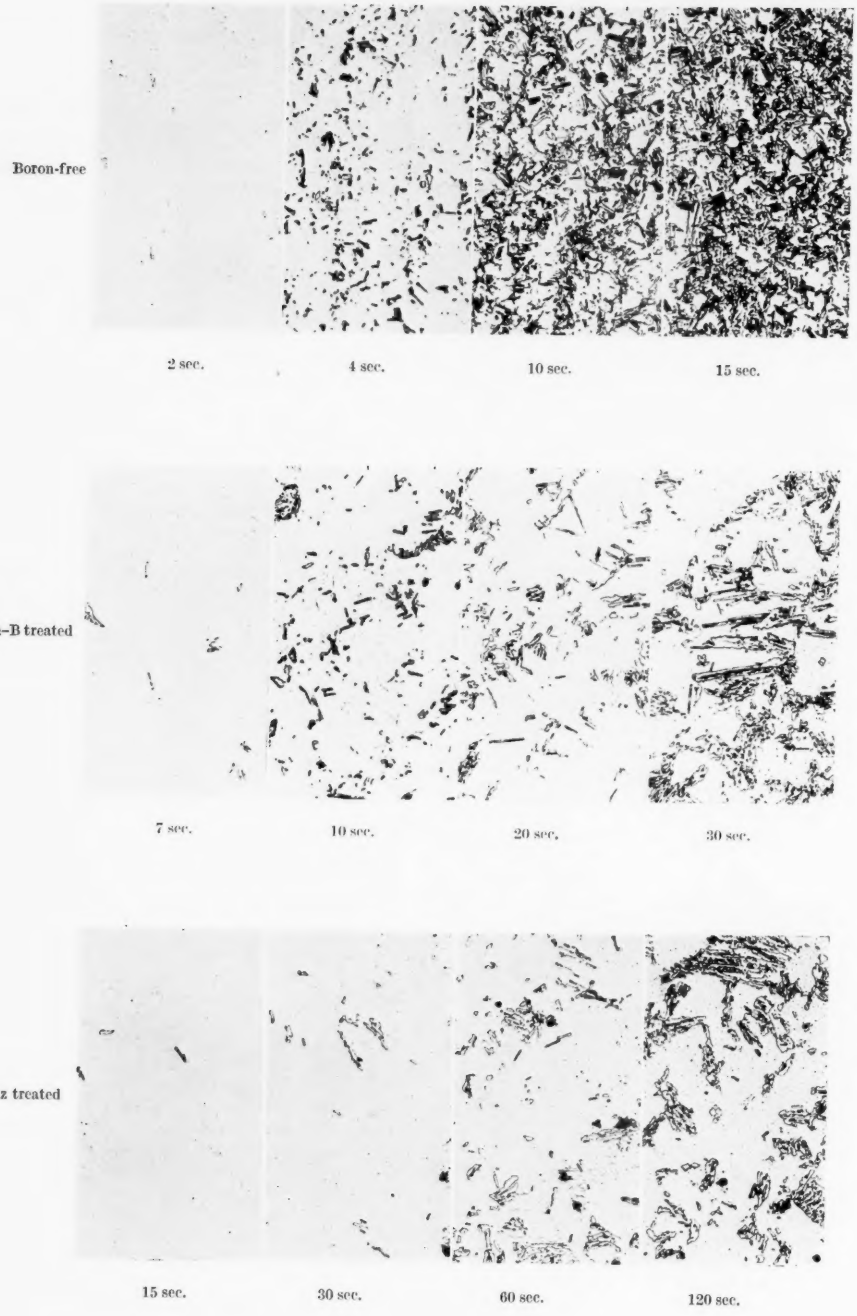

Fig. L—Microstructures of boron-free and boron-treated low-alloy steel partially isothermally transformed at 500° C. × 200

it has been found that boron steels which have been selected on the basis of duplicating the hardenability of the previous standard steel have worked satisfactorily for heat-treated parts.

Figure 3 shows that approximately 700,000 short tons of boron steel were successfully used in the United States in 1952. Relatively little difficulty was experienced with heat-treated parts. The general reports were that response to forging, welding, machining, and heat-treating operations was largely similar to that of the previous steels, although the boron steels are inclined to be lower in hardenability at the slowly-cooled end of the Jominy specimen, which means that the normalized hardness might be lower than that of the previous steel, and consequently the machinability somewhat better. We were glad to have this confirmed by one of the contributors.

Some difficulty has been encountered in case-hardening boron steels, as the boron increases the hardenability of the core much more than that of the case. This has led to some troubles with distortion, but in spite of this a considerable tonnage of boron steels was successfully used for case-hardened gears. At the time of writing (December, 1953), however, the trend seems to be away from boron steels for such gears.

In 1952 my Company used about 80,000 short tons of boron steel (*see* Fig. 3), which was equivalent to about 73% of their total alloy requirements. Again no difficulty was experienced with heat-treated parts, but some troubles were encountered in case-hardening. We are still continuing the use of the boron steels for our heat-treated parts, with a saving of about $500,000 per year on the basis of 1952 production.

Many of the passenger-car producers have used low-alloy steels, such as 1300, 4000, and 5100, which have a lower hardenability than that of the 8600, or the substitute boron steels which have been created. Steels in this range of hardenability could be produced within the government restrictions of 0·40–0·60% Ni and 0·15% Mo maximum. It was not necessary or economical to use the higher-hardenability 80*B*00, 81*B*00, or 50*B*00 boron steels for such applications. Consequently, the boron steels were used largely by truck, tractor, and heavy-machine industry, which had previously used medium- and high-alloy steels. This is the reason why only 8–9% of boron steels were used in the United States.

The question has been asked whether the successful performance of boron steels for our heat-treated parts was due to over-design. Figure 14 shows the results of fatigue tests on 3⅛-in. dia. tractor axles; these were made on the finished axles, not arbitrary specimens. The fatigue strength of the 50*B*50 boron steel was as good as that of previous steels (including 4140, 8640, etc.). Similarly, fatigue tests on springs have shown as high an endurance limit in fatigue for boron steels as for the other types of steels, heat-treated for the same Brinell hardness. This was true despite Mr. Wilcock's observation that the notched-bar impact strength in the centre of a 1⅛-in. dia. heat-treated bar of boron steel would probably be lower than that of other types of alloy steel. We agree with his findings in regard to impact strength, but we believe that both the torsion fatigue results and two years of actual service indicate that the notched-bar impact properties of the centre of the axle are not significant in determining the service performance of the final part.

This shows up some major differences in American and European practices. In the American automotive industry a great deal of attention is being given to results of fatigue and the simulated service tests of finished parts, such as crankshafts, axles, steering knuckles, gears, etc. In many cases it has been found that the results of these tests do not agree with predictions which might be made from tensile or impact tests on specimens from the centre or ½-radius position of heat-treated bars. We believe that metallurgists in all countries are fundamentally interested in guaranteeing successful performance of the final part in service; it is only a question of what is the best criterion for predicting success or failure in service.

It seems to be generally agreed that the strength at each point over the cross-section of any part must be higher than the stress to be encountered; as hardness is proportional to strength, this means that a certain minimum cross-sectional hardness pattern must be produced. The next requirement is that the part must have sufficient toughness to prevent brittle types of failure. Lastly, the part must not have a type and degree of internal stress which will produce failure.

Jominy has reported that, in automotive parts, the highest-stressed area seldom occurs deeper than one-quarter of the distance from surface to centre (¾-radius). Consequently, demanding high mechanical properties at the centre, or even ½-radius, of large sections is unnecessary, and frequently leads to specifying a higher alloy content than is actually needed to guarantee that the strength at all points on the cross-section is greater than the stress to be encountered.

It has also been found that by selecting steel on the basis of hardenability it is possible to guarantee that the *usual* heat-treating practice will produce the desired cross-sectional hardness in the as-quenched condition and, similarly, that the hardness will be adequate to permit a reasonable draw or tempering operation. It has not been necessary to use special heat-treating practice with the boron steels to produce these results; they respond similarly to the previous steels, if they are selected on the basis of equivalent hardenability.

There has been a trend in the U.S.A. towards the use of induction hardening, torch hardening, and time quenching of 1045 and other plain carbon steels, to produce medium and large size shafts which give performance superior to that of the deep-hardened alloy steels, such as 4140, 8640, 4340, etc. Specimens taken from the centre, or ½-radius, positions of these parts would predict inferior performance.

Probably the greatest difference of opinion between many American metallurgists and many European metallurgists is in regard to toughness and the methods of determining it (note that we have said 'many' and not 'all,' as we are conscious of the fact that not all American metallurgists agree on any one point, and we suspect that some differences of opinion may exist in other countries also). It seems to be generally agreed that there must be sufficient toughness to prevent a brittle type of failure. Brittle failures of well-designed automotive parts are not common in the U.S.A.; most automotive failures of steel parts seem to originate in a ductile manner, and proceed as fatigue failures. Steels that have been used in the past for automotive parts apparently produce adequate toughness to prevent brittle failures. The same has been true of the boron steels. In general, we would agree with Mr. Wilcock's findings that boron steels which are heat-treated to a high percentage of martensite will be as tough, and possibly tougher, than other alloy steels, but may not be as tough in a slack-quenched or semi-hardened condition.

Many of the contributors placed great emphasis upon the notched-bar impact properties of the centres of large sections of heat-treated boron steels. We believe that this test would be better described as a notch toughness test. The term 'impact' frequently leads to the idea

that the results have something to do with a production part being able to stand shock. This may be far from the truth. In Fig. 6 it can be seen that a steel which showed poor impact at zero with a 0·009-in. rad. fillet would have very good impact properties at − 40° F. with a commercial fillet of $\frac{1}{32}$ in. Figure 5 shows that springs made of 5152 steel have poor notch toughness at + 40° F., whilst production springs of this type have been used for many years at sub-zero temperatures without failure. This is probably due to the fact that a well-designed production spring does not have a notch. It is possible that the notch toughness of the centre of our $3\frac{1}{8}$-in. dia. axle of boron steel might be somewhat lower than that of another alloy steel, but the torsion fatigue or simulated service test was just as good.

Although many research papers deal with notch toughness and transition temperature tests, notched-bar impact tests are not considered generally necessary throughout the automotive industry. Those who have used them have stipulated rather low minimum requirements, with which we agree.

We agree with Dr. Allen that increasing the phosphorus content up to 0·007% will increase the tendency to brittleness. However, the steels covered by the paper were all basic O.H. steels, and the phosphorus content is specified as 0·004% max. and is usually held below 0·003%. Under these conditions we have found it safe to select new steels on the basis of duplicating hardenability, without fearing the brittleness that would be produced by 0·007% of phosphorus.

We are interested in Mr. Harris' contribution concerning the effect of boron on high-alloy steels; we have not had any specific experience in this field.

In regard to Mr. Toft's observations on the effect of boron addition on the impact properties of other alloys, we would expect that adding boron to a steel such as 8640 would increase the hardenability and consequently the percentage of martensite obtained in the centre of large sections. In general, we have found that better impact properties are obtained with higher percentages of martensite. We in the U.S.A. have been more concerned with duplicating the hardenability of a previous steel than with increasing the hardenability.

With reference to Mr. Bardgett's question concerning the practical interpretation of Jominy tests of carburized specimens, we agree that the cross-sectional hardness at any *one* location on the Jominy specimen cannot represent the hardness traverse on a case-hardened part, but readings at different locations may. Any one distance from the quenched end of the Jominy specimen represents a given speed of cooling. This does not change much over the cross-section, as the heat is abstracted from the end. Different distances from the quenched end represent different distances below the surface of the commercially quenched production part. For example, $\frac{6}{16}$ in. from the quenched end may represent the surface of the production part, $\frac{7}{16}$ in. the bottom of the case, and $\frac{8}{16}$ in. the centre of the core. The different layers on the carburized Jominy specimen are read at these distances from the quenched end.

Finally, I would like to plead for further study of the performance of finished parts, and the correlation of various laboratory tests with such performance. In my Company we are very seriously concerned with such problems. Contrary to the implications of its name, the International Harvester Company is one of the largest producers of motor trucks in the world, including heavy-duty trucks which haul enormous loads over rough roads; it is also one of the largest producers of heavy crawler tractors, which are called upon to substitute for dynamite in blasting rocks out of the sides of mountains. We have found that boron steel parts have worked as satisfactorily as parts made from other alloy steels for such applications.

Mr. **R. Wilcock** said in reply: Several speakers have quoted results of their own tests, and in general these have confirmed my principal findings that, although hardenability is improved by boron treatment, there is a deterioration in impact properties unless the section is fully quenched. Several speakers have emphasized the difference between the American and British approach to the question of the importance of impact properties. Another important difference in this respect is caused by the different pattern in industry in the two countries. Whereas an American steelmaker can, as it were, tailor-make a steel for one particular application, his British counterpart makes a cast which may go to ten different consumers; although he intends the boron steel for use in $1\frac{1}{2}$-in. or 1-in. dia., he may well find that his customer, in good faith, uses it in 3-in. dia. ruling section, with disastrous results on the mechanical properties.

Dr. Northcott asked why the British steel industry does not make basic open-hearth casts of 70 or 80 tons. The answer is that it does. Large open-hearth casts of a Mo–B weldable steel (Fortiweld) are regularly being made. If large casts of other grades can be absorbed in suitable applications, there is no production difficulty standing in the way of making such casts.

On the question of costs, it should not be believed that, because we use only about 1 oz. of boron per ton of steel, this ought to be had for nothing. But in addition to the boron, probably 1 lb. of titanium and 1 lb. of aluminium extra are needed, and this adds up to an appreciable amount when reckoned in the percentage of other alloys which could be obtained for the same cost. Some of the nickel or molybdenum can be obtained from scrap, and for what it costs to treat with boron we could get something like 0·4% Ni, 0·15% Mo, over 0·5% Cr, or well over 1% Mn. Even so, if a steel were being made specifically for one particular purpose many savings would be made; it should be pointed out, however, that many of these savings could be made by not using boron at all, but by using a steel just good enough for one particular application. This would involve a multiplicity of very narrow specifications.

The Chairman: I am glad that Mr. Middleton referred to our appreciation of the work that has been done in the U.S.A. on this important problem. I should like to propose a hearty vote of thanks to the authors.

The vote of thanks was carried by acclamation, and the proceedings then terminated.

REFERENCES

1. T. A. FRISCHMAN: "Alloy Conservation in Steels for Truck Axle Parts," *Eaton Engineering Forum*, 1952, vol. 13, No. 5, Sept.
2. A. J. K. HONEYMAN: *Metal Treatment*, 1944, vol. 11, p. 209, Fig. 3.
3. J. GLEN: Symposium on the Hardenability of Steel. Special Report No. 36, pp. 356–400: 1946, London, The Iron and Steel Institute.
4. C. L. M. COTTRELL: *J. Iron Steel Inst.*, 1953, vol. 174, pp. 17–24.
5. T. G. DIGGES and F. M. REINHART: *J. Res. Nat. Bur. Stand.*, 1947, vol. 39, pp. 67–131.
6. R. A. GRANGE, W. B. SEENS, W. S. HOLT, and T. M. GARVEY: *Trans. Amer. Soc. Metals*, 1950, vol. 42, pp. 75–111.
7. P. R. WRAY: *Steel*, 1951, vol. 129, No. 1, July 2nd, pp. 78–82.

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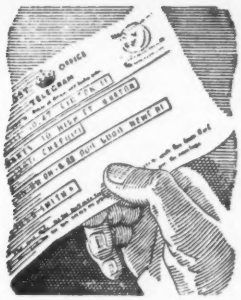

ANNOUNCEMENTS AND NEWS OF SCIENCE AND INDUSTRY

THE IRON AND STEEL INSTITUTE

Annual General Meeting, 1954

The **Annual General Meeting** of The Iron and Steel Institute will be held at 4 Grosvenor Gardens, London, S.W.1, on **Wednesday and Thursday, 26th and 27th May, 1954.** The Annual Dinner will be held on the evening of 26th May. Details of the programme will be given in a later issue of the *Journal.*

Special Meeting in Sweden, 1954

At the kind invitation of the Board of Jernkontoret a special meeting of The Iron and Steel Institute will be held in Sweden from **5th to 19th June, 1954.**

For the convenience of those members who do not wish to attend over the whole period, the meeting has been divided into three parts.

The proposed programme is as follows:

First Part (Saturday, 5th June, to Thursday, 10th June)

Preliminary to the main meeting, visits to works and other places of interest will be made at Gothenburg, Västerås, and Surahammar. (Departure from London by steamer on 5th June; Gothenburg 7th June and 8th June; Västerås etc. 9th June; arrival Stockholm evening 9th June.)

Second Part (Thursday, 10th June, to Saturday, 12th June)

The main meeting in Stockholm will begin with a reception at Jernkontoret on the evening of 10th June. There will be a technical session on the morning of Saturday, 12th June. Visits to works and places of interest will be arranged and evening entertainments provided. (Departure from London by steamer 8th June; arrival at Stockholm, *via* Gothenburg, afternoon 10th June.)

Note: Members wishing to return by steamer *via* Gothenburg can leave Stockholm on Tuesday, 15th June.

Arrangements will be made for those who prefer to travel by air to join the above parties at either Gothenburg or Stockholm; they can return from Stockholm on Sunday, 13th June, or later.

Third Part (Sunday, 13th June, to Saturday, 19th June)

After the main meeting, there will be three alternative excursions, by special sleeping car trains, to Swedish industrial centres. The first two of these, to Värmland (Hellefors, Uddeholm, Bofors, etc.) and Dalecarlia (Sandviken, Domnarfvet, Falun, Fagersta, etc.), will leave Stockholm on the evening of Sunday, 13th June, and return to Stockholm on Thursday, 17th June, or to Gothenburg for connection to steamer arriving London on 19th June. The third excursion, to Lappland (Narvik, Kiruna, Gällivare, Lulea, etc.), will leave Stockholm on the morning of Sunday, 13th June, and return to Stockholm on Friday, 18th June, to connect with the steamer leaving Gothenburg on 19th June and arriving London on the morning of 21st June.

Note: Facilities will be provided for those who wish to stay longer. Arrangements will be made for those who prefer to return by air after the excursions.

Full particulars of the programme will be given in later issues of the *Journal* and a circular will be sent to all members early in the new year.

Autumn General Meeting

The **Autumn General Meeting** of The Iron and Steel Institute will be held in London on **Wednesday and Thursday, 24th and 25th November, 1954.**

NEWS OF MEMBERS

▶ Mr. F. H. Brooks has been appointed Technical Director of the Wellman Smith Owen Engineering Corporation Ltd., and Mr. C. Brooks succeeds him as General Works Manager.

▶ Sir Henry Clay, Bt., Mr. G. N. Hewett, and Mr. T. B. Rolls have become partners in the firm of McLellan and Partners, on the retirement of Mr. A. H. Human, C.B.E.

▶ Mr. A. W. Daniels has been appointed Managing Director of Holdan Engineers (London) Ltd.

▶ Mr. V. W. Eldred has been awarded the degree of Ph.D. of the University of Cambridge, and has taken up an appointment in the Nelson Research Laboratories of the English Electric Co.

▶ Mr. P. Feltham has been awarded the degree of Ph.D.

▶ Mr. H. J. G. Goyns has been elected Vice-President of the Institution of Production Engineers (South African Branch).

▶ Mr. M. A. Grigg has left the U.K. for South Africa, where he has joined Matte Smelters (Pty.) Ltd.

▶ Mr. L. H. Keys has emigrated to Australia, where he has taken up an appointment in the Defence Research Laboratories, Maribyrnong, Victoria.

▶ Mr. D. I. King has been appointed Editor of *The British Steelmaker.* Mr. G. R. Lewington remains Managing Editor.

▶ Dr. A. H. Leckie has left the British Iron and Steel Research Association to take up an appointment as Technical Officer to the Iron and Steel Board.

▶ Mr. R. Mansford has left Dorman Long and Co., Ltd., to join Metallisation Ltd., Dudley.

▶ Mr. J. E. McLennan has left the U.K. for Australia to take up an appointment in the Aeronautical Research Laboratories, Melbourne.

▶ Mr. A. W. Postlethwaite has left the Gillette Safety Razor Co. to take up an appointment with the Raytheon Manufacturing Co., where he is attached to the Receiving Tube Division and is associated with the Germanium and Silicon Diode Section as a Senior Development Engineer.

▶ Mr. J. SAVAS has left the Packard Motor Car Co. to take up a post as metallurgist with the Alloy Research Co., Detroit, Michigan, U.S.A.

▶ Mr. HAROLD W. SECKER, O.B.E., has been appointed Chairman of Thos. W. Ward Ltd., in succession to Mr. GEORGE WOOD, C.B.E., who retired at the end of 1953. Mr. ARNOLD CARR has been appointed Assistant Managing Director, and Mr. R. F. STAGG has been appointed to the Board as a Director.

▶ Mr. J. SKINNER has been awarded the degree of B.Sc. (Eng.)(Met.) of the University of London.

▶ Mr. R. STEWARTSON has left the United Steel Companies Ltd., to join Guest Keen and Nettlefolds (South Wales) Ltd.

▶ Mr. S. F. WENNEVOLD has been appointed a Director of British Electro Metallurgical Co., Ltd., Sheffield.

▶ Mr. T. W. WHITING has left Guest, Keen and Nettlefolds, Ltd., and has left the U.K. for Canada, where he has been appointed Materials Consultant to the Royal Canadian Navy at Ottawa.

▶ Dr. F. WORMWELL, of the Chemical Research Laboratory of D.S.I.R., has been awarded the degree of D.Sc. of the University of London.

New Year Honours

▶ Mr. E. H. LEVER, Chairman and Chief Executive, Steel Company of Wales Ltd., has been created a Knight Bachelor.

▶ Mr. A. G. B. OWEN, O.B.E., Chairman and Joint Managing Director, Rubery Owen and Co., Ltd., has been awarded the C.B.E.

▶ Mr. P. O. WILLIAMS, Secretary of the Tinplate Conference, has been awarded the C.B.E.

▶ Mr. A. F. TOSH, Director and Works Manager of Summerson's Foundries Ltd., has been awarded the M.B.E.

Obituary

Mr. W. E. BENBOW, Editor of *Iron and Steel*, on 29th December, 1953, aged 53.

Sir FREDERICK MILLS, Bt., J.P., D.L., on 31st December, 1953, aged 88. Sir Frederick, formerly Chairman and Managing Director of the Ebbw Vale Works, became a Member of The Iron and Steel Institute in 1896. He was elected to the Council in 1914 and served as President for the Session 1925–26. He was made an Honorary Member in 1937.

Dr.-Ing., Dr. mont. h.c., Dr.-Ing. E. L. OTTO PETERSEN, on 27th December, 1953, aged 79. Dr. Petersen was for many years Director of the Verein Deutscher Eisenhüttenleute. He was made an Honorary Member of The Iron and Steel Institute in 1937.

Mr. F. W. SHEPHERD, Special Director and Superintendent of Production, Dorman Long and Co., Ltd., on 5th December, 1953, aged 57.

SIR ARTHUR WINDER, former General Manager and a Director of the English Steel Corporation, and an Honorary Vice-President of The Iron and Steel Institute, on 11th December, 1953, aged 78.

CONTRIBUTORS TO THE JOURNAL

James Burke, B.Eng., Ph.D.—Scientific Officer in the Research and Development Branch, Ministry of Supply, Division of Atomic Energy (Production), at Preston.

Dr. Burke was born at St. Helens, Lancashire, in 1930, and was educated at West Park Catholic Grammar School, St. Helens, and in the Faculty of Engineering, University of Liverpool, where in 1950 he was awarded the degree of B.Eng. with 1st Class Honours in Metallurgy. For the next three years he studied in the Metallurgy Department of the University, working on the graphitization reaction in iron–carbon–silicon alloys. This work was partially supported by a grant from the Andrew Carnegie Research Fund. He was awarded the degree of Ph.D. in 1953, and took up his present appointment in the same year.

C. Margaretha Hickley, Dipl. ing., Ph.D.—Technical Officer in the Research Department of Imperial Chemical Industries Ltd. (Metals Division).

C. Margaretha Hickley

Mrs. Hickley was born in Finland and graduated in metallurgy at the Technical University of Helsinki in 1948. She then took up a post in the steelworks of Oy. Vuoksenniska Ab., until in 1949 she came to England to carry out research at the University of Sheffield. She was awarded the degree of Ph.D. in 1952, and took up her present appointment in the same year.

IRON AND STEEL ENGINEERS GROUP

The **Twenty-Fifth Meeting** of the Iron and Steel Engineers Group will be held at 4 Grosvenor Gardens, London, S.W.1, on **Thursday, 18th March, 1954.** Two papers will be presented for discussion: "*Treatment of Water for the Iron and Steel Industry*," by A. J. Lamb and A. H. Waddington (Paterson Engineering Co., Ltd.); and "*Modern Industrial Boiler Plant*," by C. Ungoed (Steel Company of Wales, Ltd.). Full details of the programme will be given in the March issue of the *Journal*.

A meeting of **Junior Engineers** will be held at Buxton, Derbyshire, from 26th to 29th April, 1954. The meeting will include discussion on the commissioning of steelworks equipment and a lecture on "*The Iron and Steel Industry in the United States*." There will also be two all-day works visits.

INSTITUTE OF METALS

Awards of Medals

The Council of the Institute of Metals has made the following awards of medals for 1954:

THE INSTITUTE OF METALS (PLATINUM) MEDAL to Dr. LESLIE AITCHISON, in recognition of his services to metallurgy in industry, in education, and in public service.

THE ROSENHAIN MEDAL to Professor ALAN HOWARD COTTRELL, Professor of Physical Metallurgy at the University of Birmingham, in recognition of his outstanding contributions to knowledge in the field of physical metallurgy, with special reference to the deformation of metals.

AFFILIATED LOCAL SOCIETIES

Liverpool Metallurgical Society

The Liverpool Metallurgical Society will hold a joint meeting with the North Wales Metallurgical Society and The Iron and Steel Institute, in Liverpool, on 8th April, 1954. A paper by Mr. J. G. Wistreich on "*The Cold Working of Metals*" will be presented for discussion.

BRITISH IRON AND STEEL FEDERATION

Mr. G. H. LATHAM, Chairman and Managing Director of the Whitehead Iron and Steel Co., has been appointed President of the British Iron and Steel Federation, in succession to Sir Ellis Hunter, who has been President since 1945.

NEWS OF SCIENCE AND INDUSTRY

Conference on the Physics of Particle Size Analysis

The Institute of Physics will hold a Conference on "*The Physics of Particle Size Analysis*" at the University of Nottingham, on 6th–9th April, 1954. The Conference will be divided into four sessions: The motion of particles in fluids; the scattering of light by particles; the general phenomena encountered in particle size analysis; and the comparison of methods and the automatized methods of particle counting and size. Preprints of papers to be presented will be available beforehand and, together with a summary of the discussion, will in due course be issued as a supplement to the *British Journal of Applied Physics*. Further particulars of the Conference may be obtained from the Secretary, Institute of Physics, 47 Belgrave Square, London, S.W.1.

Dr. Eugen Piwowarsky

Dr.-Ing. habil. Eugen Piwowarsky, Director of the Foundry Institute of the Rheinisch-Westfälischen Technischen Hochschule Aachen, died suddenly at Aachen on 17th October, 1953.

Professor Kurt Rummel

Professor Dr.-Ing. rer. pol. h.c. Kurt Rummel, Founder and Director of the Institute of Fuel and Power of the Verein Deutscher Eisenhüttenleute, Düsseldorf, died on 14th December, 1953.

Mond Nickel Fellowships, 1953

The Mond Nickel Fellowships Committee announces the following awards for 1953:

Mr. J. E. BENSON (Metropolitan-Vickers Electrical Co., Ltd., Manchester) to study the technique and interpretation of results of non-destructive testing of metal components in the United Kingdom, on the Continent, and in the U.S.A. and Canada.

Mr. K. BLACKBURN (Dorman, Long and Co., Ltd., Redcar) to study hot-metal basic open-hearth practice in Great Britain, on the Continent, and in the U.S.A. and Canada, with particular reference to mixer furnace operation, refractories, instrumentation, and pitside practice.

Mr. N. B. PRATT (Broken Hill Proprietary Co., Ltd., Newcastle, N.S.W.) to study the technical and economic aspects of recent advances in the erection and operation of integrated iron and steelworks in Great Britain, on the Continent, and in the U.S.A. and Canada.

Pure Metals

In 1947 the Chemistry Research Board of the Department of Scientific and Industrial Research set up a Pure Metals Committee to investigate the position in regard to the supply of metals in a high degree of purity and to initiate any necessary work to increase the availability or purity of such materials. The Committee has purchased, through the Director of the Chemical Research Laboratories, stocks of metals of a high standard of purity, either in the form of specially prepared batches or as a safeguard against the unforeseen disappearance of a particularly good grade of material from commercial production. It is intended that these stocks shall be available for urgent research purposes, at cost plus a small handling charge.

In the following list of available materials, the figures in parentheses indicate the content of uncombined metals in the specimens. In most cases complete chemical analyses can be supplied, and the quantities available are at least on the pound scale.

Aluminium (99·99%), antimony (99·92%), bismuth (99·97%), cadmium (99·99%+), cobalt (99·95%), copper (99·99%+), gallium (99·98%+), germanium (99·99%+), indium (99·993%), iron (99·96%), lead (99·99%+), lithium (99·98%), manganese (99·8%) (remainder oxygen), nickel (99·96%+), niobium (99·7%), potassium (99·99%+), rhenium (99·99%), silicon (99·93%), sodium (99·99%+), tantalum (99·9%+), thallium (99·9%+), tin (99·99%+), titanium (99·92%), vanadium (99·7%), zinc (99·99%+), zirconium (+ hafnium) (99·75%).

Enquiries in connection with this stock should be addressed to the Secretary, Pure Metals Committee, Metallurgy Division, National Physical Laboratory, Teddington, Middlesex.

Boron and Boron Alloys

There has recently been considerable interest in the improved properties of steels and other alloys containing boron. Since before the war there have been available in this country suitable boron addition alloys for iron and steel, the non-ferrous copper and nickel base alloys, and the light metal industry, as well as smaller supplies of elemental boron. These have been produced for a number of years for direct metallurgical use as additions, and for special welding electrodes, by Murex Ltd., and much useful development work with boron has been based on these materials.

Control System for Blast-Furnace Blowers

An installation providing a reliable and flexible method of automatically regulating the volume of air supplied to blast-furnaces has recently been completed at the Rotherham Works of the Park Gate Iron and Steel Co., Ltd. A grid-controlled mercury arc rectifier is used in conjunction with standard pneumatic regulators to control the speed of the electrically-driven furnace blowers, and two blowers of the reciprocating type, operating on blast-furnace gas, have been replaced by turbo-blowers. The installation was manufactured by the General Electric Co., Ltd.

X-Ray Diffraction Equipment

A new X-ray diffraction unit, built on the system of interchangeable 'sealed-off' X-ray tubes with different anode materials and with high-transmission mica beryllium windows, has recently been introduced. The high-tension generator employs a single thoriated tungsten filament valve and the D.C. output is stabilized electronically to within 0·1%. Variations in mains voltage of up to ± 10% are compensated for automatically. Provision is also made for the X-ray tube to be used either vertically or horizontally, or, by means of a cable, away from the unit altogether. Known as the PW.1010, the equipment is manufactured by Philips Electrical Ltd.

Blooming and Slabbing Mill for Turkey

A contract of over £500,000 for the supply of a blooming and slabbing mill for the Karabuk Steel Works, Karabuk, Turkey, has been secured by the Brightside Foundry and Engineering Co., Ltd., Sheffield, against competition from Germany.

British Welding Journal

The first issue of the *British Welding Journal* was published in January, 1954. It is the official organ of the Institute of Welding and the British Welding Research Association.

Spheroidal Graphite Cast Iron

A recently issued comprehensive publication on the engineering properties of spheroidal graphite cast iron

helps to establish the position of this material relative to other cast ferrous materials. The tensile strength of the iron is about double that of a high-duty flake graphite iron, and the elastic modulus, at 25 million lb./sq. in., approaches that of steel. The publication is issued by the Mond Nickel Co., Ltd., and is No. 630 in their series.

Cathodic Protection

A book entitled "*Cathodic Protection of Pipelines and Storage Tanks*," by V. A. Pritula, has recently been published in Russia, and an English translation is now available. The book, which is fully illustrated, deals with the theory of cathodic protection and provides practical details of the method of carrying out this type of protection. Special attention has been given to methods of calculation required for various types of installations. The book is published by H.M.S.O., price 10*s*. (by post 10*s*. 6*d*.).

Institute of General Managers

A number of managing directors and general managers have recently formed a new body to be called the Institute of General Managers. The Inaugural Dinner of the new Institute was held on 20th November, 1953. Details of its formation may be obtained from the Registrar, Institute of General Managers, 86 Eccleston Square, London, S.W.1.

Change of Address

W. EDWARDS AND CO. (LONDON), LTD., have removed to Manor Royal, Crawley, Sussex (Tel. Crawley 1500).

Industrial Publications Received

▶ Articles on precision casting of the Nimonic alloys, uses of Monel in electric motors, fine bore tubing in nickel and high-nickel alloys, uses of Inconel in a submerged combustion device, exhaust valves in Nimonic 80, and fusion cutting of high-nickel alloys, are contained in "*Wiggin Nickel Alloys*," No. 22, published by Henry Wiggin and Co., Ltd.

▶ Specifications for the pretreatment of steel prior to painting, and for the application of paint, have been published by the Steel Structures Painting Council, Pittsburgh, Pa., U.S.A.

▶ Paints Incorporating Natural Rubber—British Rubber Development Board.

DIARY

1st Feb.—CLEVELAND INSTITUTION OF ENGINEERS—"*The Operation of Soaking Pits*," by J. Dodd—Cleveland Scientific and Technical Institution, Corporation Road, Middlesbrough, 6.30 P.M.

2nd Feb.—EAST MIDLANDS METALLURGICAL SOCIETY—Members' Night—Nottingham and District Technical College, Shakespeare Street, Nottingham, 7.30 P.M.

4th Feb.—STAFFORDSHIRE IRON AND STEEL INSTITUTE—"*Blast Furnaces*," by G. Hookham—Star and Garter Hotel, Wolverhampton, 7.30 P.M.

4th Feb.—LEEDS METALLURGICAL SOCIETY—"*Nucleation in Metals and Alloys*," by J. H. O. Varley—Chemistry Department, The University, Leeds, 2, 7.15 P.M.

8th Feb.—NORTH EAST METALLURGICAL SOCIETY—"*Protection of Steelwork from Atmospheric Corrosion*," by F. R. Himsworth—Cleveland Scientific and Technical Institution, Corporation Road, Middlesbrough, 7.15 P.M.

9th Feb.—MIDDLESBROUGH AND DISTRICT ASSOCIATION OF FOREMEN, ENGINEERS AND CHIEF DRAUGHTSMEN—"*Rolling Mill Practice with particular Reference to Broad Flange Beam Production*," by G. B. Thomas—Cleveland Scientific and Technical Institution, Corporation Road, Middlesbrough, 7.30 P.M.

9th Feb.—SHEFFIELD METALLURGICAL ASSOCIATION—"*High Temperature Insulation*," by G. Marsh—B.I.S.R.A., Hoyle Street, Sheffield, 7.30 P.M.

10th Feb.—NORTH WALES METALLURGICAL SOCIETY—"*Some Notes on Wire*," by R. S. Brown—County Primary School, Plymouth Street, Shotton, Chester 7.15 P.M.

11th Feb.—LIVERPOOL METALLURGICAL SOCIETY—"*Diffusion in Metals*," by A. D. Le Claire—Liverpool Engineering Society, The Temple, Dale Street, Liverpool, 7.0 P.M.

15th Feb.—CLEVELAND INSTITUTION OF ENGINEERS (Joint Meeting with The Iron and Steel Institute)—"*The New Blast-Furnace and Ancillary Equipment at Shotton*," by F. Jones—Cleveland Scientific and Technical Institution, Corporation Road, Middlesbrough, 6.30 P.M.

15th Feb.—SHEFFIELD SOCIETY OF ENGINEERS AND METALLURGISTS—"*The Future of Steel Melting*," by M. W. Thring—University Building, St. George's Square, Sheffield, 7.30 P.M.

16th Feb.—SHEFFIELD METALLURGICAL ASSOCIATION—"*Analytical Chemistry of Boron*," by H. G. Short—B.I.S.R.A., Hoyle Street, Sheffield, 7.0 P.M.

17th Feb.—NEWPORT AND DISTRICT METALLURGICAL SOCIETY—"*The Phosphate Treatment of Metals*," by H. A. Holden—Whitehead Institute, Cardiff Road, Newport, 7.0 P.M.

18th Feb.—SOCIETY OF INSTRUMENT TECHNOLOGY—"*Steelworks Instruments*," by A. E. Pearson; "*Sindri Fertiliser Factory Instruments*," by L. A. Thorman; "*The Design of Instrument Panels*," by C. M. Wells—Cleveland Scientific and Technical Institution, Corporation Road, Middlesbrough, 7.30 P.M.

23rd Feb.—SHEFFIELD METALLURGICAL ASSOCIATION (Joint meeting with the Royal Institute of Chemistry)—"25 *Years of Co-operative Research on Corrosion*," by J. C. Hudson—B.I.S.R.A., Hoyle Street, Sheffield, 7.0 P.M.

24th Feb.—STAFFORDSHIRE IRON AND STEEL INSTITUTE (Associates' Section)—"*The All-Basic Furnace*," by J. Mitton—Star and Garter Hotel, Wolverhampton, 7.30 P.M.

TRANSLATION SERVICE

(The previous announcement was made in the January, 1954, issue of the *Journal*, p. 96.)

TRANSLATIONS AVAILABLE

No. 479 (German). W. SCHEURER: "Influence of the Furnace Atmosphere on the Surface Finish of Round Ingots and Tubes." (*Stahl und Eisen*, 1952, vol. 72, July 31, pp. 935–941).

No. 480 (German). F. BLEIMANN: "The Determination of Rolling Times as a Basis for Reference—Ton Calculations in a Rolling Mill by Means of the Records of an Autographic Recorder." (*Stahl und Eisen*, 1953, vol. 73, Nov. 5, pp. 1468–1472).

No. 481 (German). F. KÖSTERS: "Quenching of Basic Bessemer Steel from the Rolling Temperature in Full-Scale Operation." (*Stahl und Eisen*, 1953, vol. 73, Oct. 8, pp. 1342–1348).

CHARGES FOR COPIES OF TRANSLATIONS—The charge for translations is £1 for the first copy and 10*s*. for each additional copy of the same translation.

ABSTRACTS AND BOOK NOTICES

IRON AND STEEL MANUFACTURE AND RELATED SUBJECTS

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MINERAL RESOURCES

Iron Ore from Nickel Ores. (*Min. J.*, 1953, **241,** Sept. 25, 361). The International Nickel Co. of Canada is undertaking the production of by-product iron ore from Sudbury (Ontario) nickel ores. The operation will ultimately yield 1,000,000 tons of high grade iron ore/year containing at least 65% iron and <2% silica.—B. G. B.

Problems of Domestic Chrome Ore Production. F. I. Bristol. (*Mines Mag.*, 1953, **43,** Aug., 26–28). The history, position and prospects of the chrome ore mining industry on the west coast of the U.S.A. are reviewed. In 1954, estimated production of concentrates will be 100,000 (short) tons per year.—K. E. J.

A Review of Chromite Resources. (*Min. J.*, 1953, **241,** Sept. 11, 295 ; Sept. 25, 362–364). The characteristics and uses of chromite ores are discussed. The Union of South Africa is the world's principal source of supply. Production in South Africa, Southern Rhodesia, the U.S.S.R., and Turkey are described.—B. G. B.

The Mineral Resources of Morvan. (*Génie Civil*, 1953, **130,** July 1, 250–252). A brief account of the geology of the Morvan district of France is given and the resources of ferrous and non-ferrous ores, fuel, and quarry stone are reviewed.

Occurrence of Manganese Ores in French Morocco. F. Reichel. (*Metall*, 1953, **7,** July, 554–555). Particulars are given of the location, quality, reserves and exploitation of manganese ores in French Morocco.—J. G. W.

Investigation of Scallon-Todd Lease Aitkin County, Minn. W. A. Grosh, J. W. Pennington, and P. A. Wasson. (*U.S. Bur. Mines Rep. Investigation* 4979, 1953, May). The work carried out indicates that there is a substantial tonnage of low grade manganese in iron formations in this area. Partial concentration can be made by simple washing. A product containing 20% Mn, 20% Fe, and 16% SiO_2 could be obtained on a commercial scale. Full details of the work carried out are given.—B. G. B.

Iron Ore: From Beach to Furnace. T. M. Rohan. (*Iron Age*, 1953, **172,** Aug. 13, 85). The new ore deposit at San Juan, Peru, has just been opened up and shipping is soon to be at the rate of 200,000 tons/month. Known reserves are over 100 million tons of over 60% iron content.—A. M. F.

Some Iron Ores of Egypt. B. W. Holman. (*Min. J.*, 1953, **240,** June 26, 766–767). Details of iron ore deposits in Egypt, notably at Asswan and Wadi Kareim, are given. Factors which make difficult the production of iron by blast furnaces in Egypt are considered.—B. G. B.

Mineral Resources and Our Economy. R. J. Lund. (*Batelle Tech. Rev.*, 1953, **2,** July, 70–75). This article surveys the present consumption of the world's mineral wealth and the likely demand in 1975. It suggests the methods which will have to be used to conserve our rapidly dwindling resources.—J. C. B.

ORES—MINING AND TREATMENT

Mechanical-Metallurgical Problems Associated with Mine Drilling Operations. T. W. Wlodek. (*Trans. Canad. Inst. Min. Met.*, 1952, **55,** 367–373). The field performances of bit, attachment, drill rod, and shank end in the drill set are analysed and their service lives are compared and causes of service failure are reviewed.—J. C. B.

Lighter Trolleys Speed Handling of Coal, Ore. H. J. King. (*Iron Age*, 1953, **172,** Aug. 20, 138–139). A brief description is given of the redesign of an ore bridge trolley to increase the handling capacity. Reduction in weight allowed an 8½-ton grab to be used instead of one of 5 tons, and the cycle time was decreased from 60 to 50 sec.—A. M. F.

Iron Ore Preparation. D. D. Howat. (*Mine Quarry Eng.*, 1953, **19,** May, 145–149). This survey pays particular attention to the preparation for the blast-furnace of the Jurassic iron ores of the Midlands. The problems of handling, processing and using this material are noted, and the best practice described. Mining, crushing, screening, drying, calcining, bedding, and agglomeration of fines are dealt with.—D. L. C. P.

A New Process of Concentration. M. D. Despujols. (*Rev. Ind. Min.*, 1953, **34,** June, 441–448). A method of gravity separation avoiding the use of fluids is described. A slightly inclined oscillating table is used with raised prisms on its surface. The theory of the method is discussed and practical information is given on the choice of linings, and the arrangement of prisms, together with a flow sheet.—A. G.

The Size Distribution of Broken Solids. J. Taylor. (*J. Inst. Fuel*, 1953, **26,** Sept., 133–138). Experiments on the breakage behaviour of lumps of a number of ores and dolomites were carried out under carefully chosen conditions of simple crushing. It was found that for both granular and non-granular materials, the 'residue' fitted the normal Gaussian distribution ; the 'complements' from solids with a non-granular structure conformed to a Gaudin type distribution. In general, the pattern of the size distribution was found to be independent of starting size and reduction ratio. The application of the results to continued crushing operations is discussed.—D. L. C. P.

Treatment of Titaniferous Magnetite Ore from Iron Mountain, Wyo. A. E. Back, C. J. Chindgren, and R. G. Peterson. (*U.S. Bureau Mines Rep. Investigations*, 4902, 1952, Aug.). In the method developed, the ore is first roasted with 15%

sodium carbonate which converts the vanadium present into a water-soluble form from which 90% of the vanadium is removed by washing. The product is then smelted in a graphite lined electric arc furnace. Substantially all the titanium is recovered as a soda slag, which, after treatment with dil. sulphuric acid, yields a product containing 75 to 80% TiO_2. About 90% of the iron is recovered as pig iron which contains minor amounts of vanadium, titanium, and sulphur.

Limestone: Unsung but Vital Rock. R. D. Raddant. (*Iron Age*, 1953, **171,** May 21, 78–79). A brief description is given of the development of the limestone deposits on the shores of Lake Huron which supplies Cleveland, Chicago, and Detroit. Strong competition has made a high degree of mechanization necessary and self unloading lake boats are used for delivery to the consumer.—A. M. F.

FUEL—PREPARATION, PROPERTIES, AND USES

Radiation and Furnace Design. G. J. Gollin. (*J. Inst. Fuel*, 1953, **26,** Sept., 151–160). Methods of heat transfer are reviewed with particular reference to gas and flame radiation, and a study is made of the application of the principles of heat transfer to furnace design in the categories : low temperature furnaces, boilers, and high temperature furnaces. Information on open-hearth and glass tank heating resulting from the Ymuiden trials is surveyed. For oil-fired, narrow-angle blast burners the momentum of the atomizing medium is the most important factor, having an optimum value. For a flame traversing a furnace it is shown that emissivity decreases uniformly with time, the flame taking 0·15–0·2 sec. to burn. For maximum radiation the flame should travel across the furnace in this time, sufficient area being left free from flame to allow the roof to radiate to the cooler charge and thus prevent overheating.—D. L. C. P.

Cleaning Various Coals in a Drum-Type Dense-Medium Pilot Plant. M. R. Geer, W. A. Olds, and H. F. Yancey. (*Min. Eng.*, 1953, **5,** July 696–705). Results are given from a pilot plant of the U.S. Bureau of Mines, using magnetite as the medium, for seven coals exhibiting a wide range in washability characteristics.—K. E. J.

Laboratory Tests of the Cleaning of Fine Coal by a D.S.M. Cyclone. T. E. Morimoto. (*Trans. Canad. Inst. Min. Met.*, 1952, **55,** 29–37). The Research Council of Alberta has made a series of tests with the D.S.M. (Dutch State Mines) cyclone washer to investigate the influence of underflow apex diameter, specific gravity of medium, pulp ratio, density distribution of the feed, and size analysis of the feed, and the results are reported.—J. C. B.

Properties of Suspensions Used in Coal Cleaning Processes. G. F. Eveson. (*J. Inst. Fuel*, 1953, **26,** Sept., 139–145). The important properties of these suspensions are reviewed, both from a practical point of view and in the light of recent research. Viscosity and rate of settling are dealt with.

The Blending of Western Coals for Production of Metallurgical Coke. J. D. Price. (*Min. Eng.*, 1953, **5,** July, 716–720). Details are given of blending experience with high-volatile coking coals (principally those of Colorado and Utah).—K. E. J.

Study of the Atmospheric Modification of Coal—Ageing. R. Wildenstein. (*Chaleur et Ind.*, 1953, **34,** Aug., 233–236). The ageing of coal, causing loss of coking properties and possible self ignition, has been investigated with the aid of a Gieseler plastometer. It was concluded that, in air, coal forms unstable oxygen compounds whose subsequent breakdown leads to the loss of fixed oxygen. At temperatures of 105–110° C. these compounds tend to become stable so that the oxygen content of the coal surface increases until combustion occurs.—A. G.

New Coking Processes. (*Organisation for European Economic Co-operation, European Productivity Agency*, Paris, July 15, 1953, DT/PR/RTP/730). This report contains: (1) A critical study of the different coking processes actually known; (2) recapitulative tables of these processes ; (3) two appendices containing short reports on carbonization processes which have not received much notice in the technical press. Under (3), A. H. Anderson refers to the travelling grate coking process as applied to a rotary lime kiln. K. Baum refers to the Convertol Process of the Bergwerksverband, and to the Baum-Panindco process. A communciation of P. Beaugrand to the IVe Congrès International du Chauffage Industrial describes the production of artificial anthracite from Carmaux coal.—R. A. R.

The Specification, Determination and Testing of Grades of Coke. (*Brit. Coke Res. Assoc. Panel No. 3, Third Report to the Research Committee*, June, 1953). The Panel reports on a study of commercial grades of coke and methods of determining their true sizes. The true size limits of a grade of coke are very different from the sizes of the screens producing the grade. Coke grades cannot therefore be properly described by using the screen sizes.—R. A. R.

Catalytic Enrichment of Industrial Gases by the Synthesis of Methane. (*Department of Scientific and Industrial Research, Fuel Research, Technical Paper No. 57*, 1953). The paper describes a laboratory-scale investigation of the catalytic enrichment of industrial gases by synthesis of methane using nickel catalysts, as well as tests with intermediate-scale plant.

The Pumping, Transmission, and Burning of Tar and Pitch. W. A. Pond. (*Iron Steel Eng.*, 1953, **30,** July, 94–97). The selection of a pump for pushing tar and pitch, of means of carrying fuel from one point to another, and the choice of equipment for effective combustion are discussed. The relative merits of tar and pitch as an open-hearth fuel are examined and preference is given to pitch.—M. D. J. B.

TEMPERATURE MEASUREMENT AND CONTROL

The Pyrostat, a New Thermoelectric Temperature Controller. A. Jagersberger. (*Osterr. Masch. Elektrowirtschaft*, 1953, **8,** Aug., 255–257). The construction, mode of operation and circuit diagram of this furnace temperature regulator are given.—L. D. H.

Automatic Controls Improve Furnace Operation. L. Walters. (*Iron Age*, 1953, **171,** Apr. 30, 108–111). The principles of automatic control of temperature by means of two-step, multi-step, and gradual methods are described. For continuous action a proportional controller may be used, working on a potentiometric system. The accuracy of control is usually about 1% of the full-scale reading. Programme controllers can be devised to produce any desired heating cycle.

REFRACTORY MATERIALS

New Silica Refractories Plant Built at Windham, Ohio. (*Iron Steel Eng.*, 1953, **30,** Aug., 124–129). The new Windham plant is described. Conventional and super-duty silica refractories are manufactured. The plant, which comprises 30 tunnel kilns, has facilities for grain size control, precision batch control, impact pressing, controlled firing temperatures, and a laboratory equipped for continuous quality control.

M. D. J. B.

Bulk Materials Handling: It Can be Done Cleanly. (*Steel*, 1953, **133,** Aug. 24, 110–112). The automatic equipment and highly mechanized materials handling systems described make up a clean and orderly silica brick plant.—D. L. C. P.

A Theory of Water: The Clay-Water System. II—Water Adsorption and Cation Exchange in Montmorillonite. E. Forslind. (*Acta Polytechn.*, 1952, No. 115). The two current conceptions regarding the origin of the base exchange capacity of montmorillonite are discussed, and the objections raised against the theory of Edelman and Favejee are subjected to critical analysis. The assumption of trans-co-ordinated silica tetrahedra in the clay crystal surfaces lead to conditions of interaction between the clay crystal and the adsorbed water, regarded as a crystalline phrase, which give rise to a charge transfer that corresponds to the observed exchange capacity of montmorillonite. The applicability of the author's theory to the Hofmann structure is briefly considered.—J. C. B.

Basic Lining of High-Frequency Induction Furnaces. F. Sellner. (*Osterr. Masch. Elektrowirtschaft*, 1953, **8,** Jan., 8–12). The necessary qualities of a basic rammed lining for an induction furnace are described, and the compositions, grading and other characteristics of burnt magnesia for this purpose are discussed in detail. The technique of ramming is outlined, and a typical heat in a 2·5-ton magnesia-lined furnace is followed.—L. D. H.

New Types of Refractories. G. Bieler. (*Silicates Indust.*, 1953, **18,** June, 234–236). An account is given of the properties, preparation, uses, and potential fields of application of new types of refractories, particularly sulphides, such as BaS; mixtures of the sulphides of thorium and cerium ; carbides of tungsten, tantalum and cerium ; borides of zinc, tantalum, tungsten, cerium ; and nitrides of boron and uranium. Silicides have also been studied. Most of the sulphides considered are suitable for steel-melting crucibles, and these require no inert or protecting atmosphere in operation.—P. F.

BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

Investigation to Determine the Reducibility of Iron Ore. C. Rikar. (*Stahl u. Eisen*, 1953, **73**, Aug. 13, 1094–1101). Reducibility tests have been carried out on three Jugoslavian ores using carbon monoxide in a tube furnace, charcoal in an alundum crucible, and Ruhr coke and lignite coke in an experimental blast-furnace with a hearth diameter of 300 mm. Results are plotted as reduction/temperature curves. It is claimed that the experimental blast-furnace is of great value in assessing the suitability of ores and sinters for smelting in small to medium sized blast-furnaces, provided that the greater coke consumption is taken into consideration. The advantages of the furnace and the rapidity with which results can be obtained are discussed.—J. P.

Plant Experiences and Economic Results from Blast-Furnace Operation with Increased Hearth Load. H. Kahlhöfer and R. Ehrhardt. (*Stahl u. Eisen*, 1953, **73**, July 30, 1014–1022). The effect of hearth loading between 700 and 1200 kg./sq. m./hr. on the operation of the four blast-furnaces at the Mannesmann works has been investigated. Within this range, an increase in hearth loading had no detrimental effect on furnace working. It caused a reduction in the amount of indirect reduction and an increase in gas temperature and therefore an inferior thermal utilization of the coke and an increase in the consumption thereof. A cost comparison shows that increased hearth loading is worthwhile because it increases production and reduces working costs.—J. P.

Rehabilitating a Blast Furnace and Associated Equipment. W. C. Daniels. (*Iron Steel Eng.*, 1953, **30**, July. 57–65). A detailed description is given of a 25-ft. hearth blast-furnace rebuild at the works of the Bethlehem Steel Co., which took only 30 days. This time was achieved by furnishing the necessary craftsman with proper equipment, good planning, and providing adequate materials-handling facilities.

Construction of Low-Shaft Blast Furnace Nears Completion. (*Iron Age*, 1953, **171**, May 21, 133–137). Details of the plant layout and construction of the low-shaft furnace at Liège, Belgium, are given. The furnace is of oval construction and about 36 ft. high. It is designed to use a maximum oxygen enrichment of the blast of 50% and will have an output of 60–100 tons of iron per day.—A. M. F.

The Mechanized Erection of Blast-Furnace Plant Using Large Reinforced Rings. N. A. Boloban. (*Mechanisation of Heavy Work* (Moscow), 1949, **6**, (6) 14–15 ; *Met. u. Giesserei Techn.*, 1953, **3**, July, 297–299). Details are given of the erection, in only five months of the No. 1 blast-furnace at the Krivoirog steelworks. Much time was saved by prefabricating the stack in rings complete with lining and shell plate, and bringing the rings within the radius of the erecting crane.

Controlling Blast Furnace Air Supplies. (*Elect. Times*, 1953, **124**, Nov. 12, 903–904). **Control System for Motor-Driven Blast-Furnace Turbo-Blowers.** (*Engineer*, 1953, **196**, Nov. 6, 590–591). Two electrically driven blowers supply air to the blast-furnaces of the Park Gate Iron and Steel Co., Ltd. Each can deliver a maximum of 30,000 cu. ft./min. of free air at 10 lb./sq. in. gauge. Each is driven through reduction gears by a 2200 h.p. D.C. motor for which current at 750 V. is supplied by mercury-arc rectifiers. The arrangement of rectifiers and Askania regulators is explained.

TREATMENT AND USE OF SLAGS

Utilization of Waste Products of the Steel Industry. S. N. Sircar. (*Tisco Rev.*, 1953, **21**, Jan., 6–17). A summary is given of the waste products obtained from an integrated steel plant. The author then considers each of the waste products as produced by the Tata Iron and Steel Company in India, and discusses the uses to which they might be put. Results likely to be obtained from them and a few results from laboratory experiments are given. Nothing has yet been done on a large scale except the use of blast-furnace slag for road-making material.—J. C. B.

The Mechanism of the Viscosity of Blast-Furnace Slags. A. M. Chernyshev, L. M. Tsylev, and A. V. Rudneva. (*Izvestiya Akademii Nauk S.S.S.R., Otdelenie Tekhnicheskikh Nauk*, 1953, (7), 1044–1057). [In Russian]. On the basis of the ionic theory of slags a theoretical interpretation of changes in the viscosity of slags with changes in their chemical composition is attempted. It is concluded that the viscosity of a homogenous liquid slag is governed mainly by the size of silicate anions : The greater the size of silicate anions and the concentration of large silicate aggregates, the stronger is the interlocking of the individual slag layers. The size of the complex silicate anions depends on the ratio of the number of oxygen atoms to the number of silicon atoms in the slag. The larger this ratio is, the smaller are the silicate aggregates and *vice versa*. Therefore, with increasing concentration in the slag of CaO, MgO, TiO_2, MnO, FeO, and Na_2O, *i.e.*, oxides which do not form complex aggregates in a liquid slag, the viscosity of the slag is decreased because of the increase in the oxygen/silicon ratio.—V. G.

Solubility Tests on Blast-Furnace Slags. J. Endell. (*Arch. Eisenhüttenwesen*, 1953, **24**, July–Aug., 281–284). Solubility tests have been carried out on blast-furnace slags with acid ammonium citrate as solvent. Shaking and percolation tests have shown that the basic constituents dissolve readily in this medium which thus appears suitable for laboratory assessment of the neutralizing power of lime manuring materials.

A Factory at Volta Redonda for Cement from Blast-Furnace Slag. L. German. (*Eng. Min. e Met.*, 1951, **16**, Sept.–Oct., 175–176). [In Spanish].

PRODUCTION OF STEEL

Materials Balance and Comparison of the Economic Efficiency of Different Steelmaking Processes. E. Krebs and R. Görgen. (*Stahl u. Eisen*, 1953, **73**, Aug. 13, 1081–1094). A method is demonstrated and illustrated by several examples by which the cost of producing steel by a variety of processes (open-hearth, converter plus open-hearth, converter plus electric arc, oxygen-blown converter, duplex converter plus open-hearth, and duplex converter plus electric arc) may be calculated.—J. P.

Open Hearth Challenged as More Carbon Steel Goes Electric. A. G. Gray. (*Steel*, 1953, **133**, Aug. 24, 84–88, 108). **Electric Furnace Challenges Open Hearth.** (*Iron Steel Eng.*, 1953, **30**, Aug., 130–133). Data on the extent of carbon steel production in electric furnaces in U.S.A. are given; recently 52% of electric furnace output has been carbon steel. A survey is made of a Battelle report comparing the cost of electric and open-hearth carbon steel ; on these figures the production cost per ton of ingots is less for the electric furnace on cold metal practice, but more with hot metal. Details are given of the two new 200-ton top-charge Héroult electric furnaces.

Increasing Basic-Bessemer Steel Production by Improving Converter-Bottom Life. K. Schröder. (*Met. u. Giesserei Techn.*, 1953, **3**, Aug., 315–323). Attention is drawn to the effect upon production of the durability of converter linings. A list of the most important sources of dolomite in Eastern Germany is given. The essential characteristics of burnt dolomite, coal tar, and basic-Bessemer pig iron are outlined. Experiments with other materials are discussed, and data are given for mechanical and chemical wear of bottom and lining. The production of tar dolomite is briefly described.—L. J. L.

Development, Practice, and Equipment of the Oxygen-Blowing Steel Plant of Vöest. H. Trenkler. (*Österr. Masch. Electrowirtschaft*, 1953, **8**, Jan., 4–6). The author, after outlining the reasons for the installation of the new plant at Linz, describes the development and principles of the ' LD ' (Linze Düse = Linz nozzle) process of top-blowing with commercially pure oxygen.—L. D. H.

Open Hearth Furnace Design and Availability. G. R. Bashforth. (*Brit. Steelmaker*, 1953, **19**, Sept., 541–549). The author first discusses melting-shop layout and enumerates some of the more important features. Open-hearth furnace design, with particular reference to the port-ends, is next considered, and descriptions are given of the conventional, Venturi, Terni, Maerz, and single-uptake designs. Finally, the factors affecting furnace availability are briefly discussed.—G. F.

Regenerator Efficiency and Air Preheat in the Open Hearth. B. M. Larsen. (*Amer. Iron Steel Inst. Preprint* P272, Oct. 21, 1953). The data presented are based on tests for regenerator efficiency made on three open-hearth furnaces representing the extremes of present operating practices. In addition, extensive calculations by an electrical analogy method produced reasonably quantitative data on regenerator efficiency and these agreed closely with the direct results of the furnace tests.—J. C. B.

How to Curb Heat Losses from Cooling Units. A. M. Frankau. (*Steel*, 1953, **133**, Sept. 21, 138–140). Figures and comments on the amount of heat carried away by the cooling water from open-hearth furnaces are given. Losses up to

27½ %of the total heat input are reported, representing over 40% heat input for the cooling units. It is suggested that refractory protection should be given were possible.—D. L. C. P.

Hot-Water Cooling of Open-Hearth Furnaces. A. Harnisch. (*Stahl u. Eisen*, 1953, **73**, July 30, 1026–1028). A system is described in which hot water is pumped through cooling tubes in door frames and gas up-takes of a 60-ton open-hearth furnace. The resulting steam-water mixture is separated, the steam being fed to the works system and the water, with necessary topping up, pumped back to the furnace cooling tubes. The advantages of the system are: (1) Reduced circulation of water because the cooling is effected by vaporization ; (2) lower pumping costs ; (3) high-grade steam is produced ; and (4) the possibility of using treated water and so avoiding scale deposition in the cooling tubes. Initial costs were paid off in two months. After this, the system resulted in a net saving of DM 3·12/ton of steel produced.—J. P.

Saving Water in Open-Hearth Furnace Cooling Systems. H. Lardy. (*Stahl u. Eisen*, 1953, **73**, July 30, 1028–1030). Hot cooling of door frames and the resulting production of steam saves fuel, cooling water, and electric current. The equipment is so small that it can be installed even in the most restricted areas. It does not require separate operating personnel nor does it put additional duties on existing furnace crews. Capital costs are soon written off.—J. P.

High Pressure Steam from Open-Hearth Furnace Waste Gas and Cooling Water. K. E. Poppe. (*Stahl u. Eisen*, 1953, **73**, July 30, 1030–1035). The use of a waste heat boiler and hot water cooling and the necessary equipment are described. Water is circulated from a high pressure steam vessel, through a waste heat boiler (heated by the waste gases from the furnace) and then through high-pressure door-frame cooling tubes and back to the steam vessel. The steam separated here is fed to turbines for an air compressor and, if desired, for electricity production. Water coming from gas port cooling tubes is passed through a heat exchanger to take heat from the waste steam from the turbine. The heated water is used for space heating and then circulated again through the gas ports. The condensed steam from the heat exchanger is fed back to the door-frame cooling system. The equipment is safe and reliable and saves DM 7–10/ton of steel produced, and capital costs can be written off in less than a year.—J. P.

Life of All-Basic Open-Hearth Furnace Roofs in Special Steel Practice. O. Krifka and A. Schöberl. (*Radex Rundschau*, 1953, (2), Feb., 58–65). The advantages of using chrome-magnesite linings are described and details of furnace construction are discussed. Performances of open-hearth furnaces with acid and basic linings are compared and extensive data are quoted. Various improvements are described which increase the all-basic furnace performance to 2000 heats and keep the heat consumption per ton of steel constant almost to the end of the furnace campaign.—E. C.

Some Experience with an All-Basic Tilting Furnace. J. Mitton. (*J. Jun. Inst. Eng.*, 1953, **63**, Sept., 373–391). An account is given of the design and running of a 60-ton oil-fired, all-basic tilting furnace at the Bilston works of Stewarts and Lloyds Ltd. It is claimed that British chrome-magnesite bricks are as good as the Radex E brick in service. Details are given of the instrumentation which, combined with oil firing, is claimed to be ideal for operation of a small tilting furnace.—A. D. H.

The Manufacture and Use of Boron Steels in the U.S.A. L. J. Rohl. (*J. Iron Steel Inst.*, 1954, **176**, Feb., 173–180. [This issue].

Present State of Development of German Open-Hearth Furnaces. A. Mund and C. Kreutzer. (*Stahl u. Eisen*, 1953, **73**, Sept. 10, 1201–1212). The means employed in Germany to improve the efficiency and life of open-hearth furnaces are briefly described. The designs of furnaces, checkers and roofs, methods of producing draught, carburizing fuel gases, use of oxygen for combustion and for refining, and the use of pre-melted scrap are discussed. The success of measures being applied is reflected in the present increased output of German furnaces.—J. P.

Operating and Technical Practices in the European Iron and Steel Industry. W. C. Bell. (*Brit. Steelmaker*, 1953, **19**, Aug., 474–485 ; Sept., 524–533 ; *Amer. Iron Steel Inst. Yearbook*, 1953, 104–159 ; *Indust. Heating*, 1953, **20**, Aug., 1527–1538, 1634). In this review of European operating and technical practices, the author deals first with ore mining and preparation of the blast-furnace burden, mentioning vacuum extrusion, pelletizing and beneficiation of low-grade ores. Coke-oven practice and efficient treatment are next considered, and present-day techniques and trends in ironmaking are outlined. In considering European steelmaking practice, particular reference is made to the development and future possibilities of the basic-Bessemer process and to the new flexibility given to the process by oxygen and oxygen–steam blowing. The possibility of a new process combining Bessemer and open-hearth refining in one vessel is mentioned.—G. F.

Operational Studies in an Open-Hearth Steelworks. F. W. Tonnius. (*Stahl u. Eisen*, 1953, **73**, July 30, 1042–1046). Operational studies indicate that, in certain instances, charging time depends on the way in which the charging crane is used. A greater part of the lost time can be avoided by supervision and operational planning. It is shown that in this way two cranes, the capacity of which appeared to be scarcely sufficient for three furnaces, can supply four furnaces without increasing charging times.—J. P.

Slag and the Refining of steel in the Open-Hearth Process. III—Sulphur. E. Bučko. (*Prace Instytutow Ministerstwa Hutnictwa*, 1953, **5**, (3), 143–158). [In Polish]. It has been determined by the method of statistical analysis of results obtained from two steelworks, that the principal factor affecting the distribution of sulphur between slag and metal $N_S = (S)/[S]$ in the basic open-hearth process is the slag basicity $V = CaO/SiO_2$. Other factors, such as FeO in the slag and the carbon and manganese contents of the metal have a definite but minor influence. N_S will remain below 10 even at a high slag basicity V of 3 to 5. Results obtained using Herasimenko's and Speight's formulæ were in agreement with the author's observations, whilst the formulæ of Schenk, Darken, and Larsen give higher values.—V. G.

Thermodynamic Observations on Deoxidizing with Aluminium. W. Gilde. (*Met. u. Giessereitechn*, 1953, **3**, Aug., 324–328). Some approximations for the calculation of free enthalpy are discussed. The simple first approximation would appear to be satisfactory in practice. Enthalpy-temperature functions have been worked out for the reactions of aluminium with oxygen, sulphur, nitrogen, and carbon. It is shown that the values obtained by analysis correspond with the equilibrium obtaining near the freezing point.

Experiences with Oxygen Control in Open-Hearth Flue Gases. F. P. Hubbell. (*Iron Steel Eng.*, 1953, **30**, Aug., 53–56). A description is given of an installation for continuous sampling and control of oxygen content of furnace flue gases made on a 225-ton basic open-hearth furnace installed at the Homestead Works of the United States Steel Corp.—M. D. J. B.

The Activities of Carbon and Oxygen in the System Liquid-Iron/Carbon/Oxygen. A. M. Samarin. and R A. Karasev. (*Izvestiya Akademii Nauk S.S.S.R., Otdelenie Tekhnicheskikh Nauk*, 1953, (8), 1130–1136). [In Russian]. The results of Marshall and Chipman on the dependence on concentration of the activity coefficients of carbon and oxygen in molten iron are discussed. It is concluded that in molten iron–carbon alloys containing up to 1% of carbon the activity coefficients of carbon and oxygen are constant.—V. G.

American Arc Furnace Practice. W. B. Wallis. (*Birlec Ltd.*, Sept. 30, 1953. Pamphlet : *Foundry Trade J.*, 1953, **95**, Oct. 22, 507–511 ; *Brit. Steelmaker*, 1953, **19**, Nov., 670–672 ; *Iron and Steel*, 1953, **26**, Nov., 512–523 ; *Iron Coal Trades Rev.*, 1953, **167**, Oct. 9, 827–833 ; *Mining J.*, 1953, **241**, Oct. 9, 416–417). The development from 1906 to the present time is traced. In steel foundries, nearly all arc furnaces are under 20 tons, but two 50-ton furnaces will soon be in operation at a large foundry. Progress in electric furnace size and transformer ratings are shown by graphs.—R. A. R.

Excellent Experience with On-Load Tap Changers for Electric Steel Melting Furnaces. G. Keller. (*Brown Boveri Rev.*, 1952, **39**, Nov.–Dec., 454–455). A brief note records the satisfactory performance of an on-load tap changer used on a Swiss 15-ton furnace.—A. D. H.

Modern Arc Melting Furnace Plant Features Swivelling Roof in Rapid Top Charging. R. Lambert. (*Brown Boveri Rev.*, 1952, **39**, Nov.–Dec., 429–435). A 20-ton capacity arc melting furnace in a Belgian works is described with special attention to the electro-hydraulic electrode regulating system.

Some Factors Affecting the Wear of Graphite Electrodes in the Electric-Arc Furnace. Electric Process Sub-Committee. (*J. Iron Steel Inst.*, 1954, **176**, Feb., 159–165). [This issue].

Gravity Helps Double Stainless Output. (*Steel*, 1953, **133**, Sept. 14, 108–110). A brief description is given of the new

electric melting shop and rolling mill at the Watervliet plant of the Allegheny Steel Corp. There are three American Bridge Héroult furnaces, two of 10,000 kVA. each, and one of 75 kVA. In the rolling mill there are a 22-in. two-high reversing blooming mill and an 18-in. two-high non-reversing stand.

Arc Furnace Electric Control: Advantages of Hydraulic System. (*Brit. Steelmaker*, 1953, **19,** Sept., 556–557). Methods of controlling electric arc furnace electrode movement are briefly discussed. Particular attention is given to the Tagliaferri method of hydraulic furnace control, constituting motive power for tilting the furnace, swinging the roof, raising the door, and operating the electrodes. The advantages of the system are discussed.—G. F.

Recovery of High-Speed Steel Scrap by Melting in an Electric Arc Furnace. K. Radzicki. (*Prace Instytutow Ministerstwa Hutnictwa*, 1953, **5,** (3), 113–118). [In Polish]. The literature on recovering alloying components in high-grade alloy steel scrap is reviewed and the results of experimental heats after various degrees of oxidation are reported. Best results were obtained when melting was carried out using 95% high-speed steel scrap and 5% scale.—V. G.

Steel Melting in the High Frequency Furnace. A. Mettam. (*E.S.C. News*, 1953, **7,** Aug., 2–7). The development of the high-frequency induction furnace is reported with special reference to the three 3-ton and three 1-ton units at the River Don works of the English Steel Corp. The production of the monolithic lining is briefly given and followed by the advantages of steel melting in this type of furnace.—J. C. B.

Continuous Casting of Iron. A. N. Mjassojedow and I. R. Dudnic. (*Foundry Production* (Moscow), 1952, (11), 2–5 ; *Met. u. Giesserei Techn.*, 1953, **3,** July, 277–280). The machine described is capable of producing billets 150 mm. in dia. and 1700 mm. long. Details of analysis of castings are given to show the satisfactory performance of the machine.

The Junghans Method of Continuous Casting of Steel. K. G. Speith and A. Bungeroth. (*Metal Treatment and Drop Forg.*, 1953, **20,** May, 197–204 ; June, 268–272). Details are given of experimental work carried out by the authors in conjunction with Dr. Junghans at the Mannesmann-Hüttenwerke A.G., Huckingen, Germany. Billets up to 10·4 in. dia. and 11·8 in. square have been continuously cast in plain carbon, rimmed, tool, and stainless steels. The process of solidification and the influence of temperature on casting speed are discussed. The importance of solidification time, its influence on the necessary height of the casting unit, and the furnace capacity to ensure an adequate supply of molten metal are also considered.—P. M. C.

Banded Structures in Steels. D. G. Walker. (*Australian Inst. Metals* : *Australasian Eng.*, 1953, May 7, 63–69). The author discusses the nature and origin of the three main classes of banding ; Carbon banding, alloy banding, and slag banding. Attention is concentrated on 'eutectoid banding' (*i.e.*, carbon banding), and its origin is discussed in terms of the three major theories : Alloy segregation, slag segregation, and dynamic segregation. (45 references).—P. M. C.

FOUNDRY PRACTICE

The Technical and Economic Basis for Non-Ferrous/Heavy-Metal Composite Castings. G. Schwietzke. (*Zeitschrift für Metallkunde*, 1953, **44,** Jan., 8–17). **Composite Castings.** G. Schwietzke. (*Metal Ind.*, 1953, **83,** July 10, 23–26; July 17, 43–45). The article surveys the technical and economic aspects of non-ferrous/ferrous composite castings, particularly the bonding of copper alloys to steel. Development of the methods is due to the desire to combine the wear resistance and bearing properties of brass, bronze, and leaded bronze with the strength of steel, and also to save non-ferrous metals. Brief details are given of no less than sixteen processes, all differing in same respect. The various methods comprise casting-on, casting-round, immersion, through-flow, and centrifugal techniques for bringing the molten non-ferrous metal into contact with the steel.—P. M. C.

A Practical Approach to Casting Design. R. J. Franck. (*Product Eng.*, 1953, **24,** July, 192–195). The author discusses and illustrates the use of experimental stress analysis and scale models to simplify and improve the design of steel castings. Stress coatings and strain gauges are used to find stress concentrations.—A. M. F.

Annealing and Melting with Low-Frequency Induction Heating. K. Fraunfelder. (*Brown Boveri Rev.*, 1952, **39,** Nov.–Dec., 410–420). The general features of low-frequency induction heating are discussed. Generally the process is inefficient for small parts. The low-frequency induction melting furnace previously used for melting brass is now used as a holding furnace for grey iron. Brief details of typical installations in German and Italian works are given.—A. D. H.

Casting Ingot Moulds. J. E. Rehder. (*Canad. Metals*, 1953, **16,** July, 30–31). A description of the new foundry plant at Hamilton, Ontario, of Canada Iron Foundries Ltd. is given. A production of 400 tons/day has been designed for the production of ingot moulds. The iron is supplied from two cupolas with a melting rate of 24 tons/hr each.—J. C. B.

How to Improve Carbon Pickup in Foundry Iron. E. Fry. (*Iron Age*, 1953, **172,** July 16, 140–141). Experiments with a new high carbon coke have shown that it is possible to get greater carbon assimilation in foundry iron. It permits the use of lower bed heights and gives faster melting with less refractory wear. The chief disadvantage is the requirement for about twice the quantity of limestone flux.—A. M. F.

Hot Metal Receivers. A. J. G. Smith. (*Engineer Foundryman*, 1953, **18,** Aug., 61). An oil-fired furnace for holding iron, fed by one or more cupolas, is described. It is claimed that foundry production may be improved by 30% by its use.

Study on the Basic Cupola Practice. I. T. Tanaka, T. Maruo, and N. Matano. (*J. Mech. Lab.*, 1953, **7,** Mar., 38–81). [In Japanese]. Metal and slag reactions, and production of nodular cast iron, were investigated in experimental heats on a small cupola. With slag basicity in the range 2·5–3·0, metal containing 4% C and 0·01% S was obtained directly. Production of nodular iron is greatly influenced by the quality and history of the charge metal in basic practice. (11 references).—K. E. J.

The Actual Blast Volume in a Cupola. M. Czyzewski and C. Podrzucki. (*Przeglad Odlewnictwa*, 1953, **3,** (7), 202–210). [In Polish]. Attention is drawn to the large blast losses in cupola operation which are seldom taken into consideration in evaluating cupola performance. For this reason the concepts of real (*pr*) and apparent (*pa*) blast volume are introduced. The ratio *pr*/*pa* is called blast efficiency, which in modern cupolas varies between 0·5 and 0·98. Literature dealing with cupola performance is critically surveyed in the light of apparent and real blast volume. The dependence of the coke consumption on the ratio $CO/(CO_2 + CO)$ and on the real blast volume is discussed for various cupola outputs using Polish coke. Methods for the approximate and exact determinations of real blast volume are described.—V. G.

Cupola Melting of Grey Iron—Some Consideration of Fluxes. D. A. Dodson. (*Canad. Metals*, 1953, **16,** June, 38–40). The production of clean cupola iron depends upon several factors. Desulphurization demands a basic slag but this slag must flux the sand, coke ash, and rust charged into the furnace. It must also prevent oxidation of the iron and its alloying metals so that a low melting point is very desirable to provide a slag blanket over the metal pool as early as possible in the operation. Magnesium and neutral calcium salts are recommended to attain this property in preference to the highly basic but destructive sodium salts.—J. C. B.

The Use of Scrap in the Cupola. (*Usine Nouvelle*, 1953, May 7, 60). The problem of charging larger percentages of steel scrap in the cupola is discussed. The main difficulty arjses from the different temperatures of fusion of iron and steel. The effects of steel on the conditions of fusion, chemical reactions, and the other constituents are outlined. Metallurgically, steel refines the grain giving an iron with a pearlitic homogeneous structure, having good castability and mechanical properties.—A. G.

The "Bondact" Process and Its Advantages for Iron and Steel Foundries. W. Krivan. (*Osterr. Masch. Elektrowirtschaft*, 1953, **8,** Jan., 18–19). The Bondact machine and process for projecting refractory material through a flexible pipe to make cupola linings, or repair old ones, are described. It is made (by Bonvillain and Ronceray, Choisy-le-Roi, Seine) in two sizes projecting 25 and 50 kg. of material per min. It is driven by a ½ h.p. motor and uses 3 cu.-m./min., the pressure in the pipeline is 6 atm. It can also be used for ladles, annealing furnaces, and kilns.—R. A. R.

Applications of Electric Heat in Grey-Iron Foundries. G. Keller. (*Brown Boveri Rev.*, 1952, **39,** Nov.–Dec., 403–409). Arc and induction furnaces are briefly described and compared with the cupola. Electric mould drying and annealing furnaces are also discussed.—A. D. H.

Production of Nodular Iron with Calcium. L. A. De Araujo. (*Quarterly of the Colorado School of Mines*, 1953, **48**, Jan., 43–67). The author gives an account of laboratory and foundry-scale tests to determine whether calcium can be used to produce nodular cast iron. Additions of calcium and calcium alloyed with silicon and copper respectively, were made. The tests were unsuccessful and it was concluded that calcium additions cannot compete with the magnesium process. A. G. H.

Formation of Nodular Graphite in Cast Irons. V. A. Altekar. (*Quarterly of the Colorado School of Mines*, 1953, **48**, Jan., 1–41). A review of previous work on the chemical and thermal factors affecting graphitization indicates that carbon and silicon profoundly influence graphitization and that hydrogen and sulphur are carbide stabilizers. The rate of cooling through the eutectic temperature is the most important thermal factor. Theories of the formation of nodular graphite are critically compared. By melting 150-g. charges of three commercial irons containing 3·1% C, 2% Si, 0·1% S, inoculated with a Mg–8%-Si, alloy and allowing them to cool slowly to between 1371° and 1121° C., and then quenching, the author suggests that the formation of nodular graphite involves an ordered decomposition of an austenite-carbide eutectic between 1190° and 1121° C.—A. D. H.

The Status, Development and the Possible Australian Future of Spheroidal Graphite Cast Irons. L. C. Bogan. (Australian Inst. Metals : *Australasian Eng.*, 1953, Mar. 7, 66–72; Apr. 7, 46–52; May 7, 42–46). A history of spheroidal graphite structures in cast iron is given, and a short account of the basic metallurgical principles for the commercial manufacture of this iron, by the magnesium process, is provided. The suitability of present day Australian hematite pig irons for the magnesium process is considered. The manufacturers of foundry irons can supply a material which is adequate to permit of the development of this iron in Australia by use of the acid cupola. However, if the highest ductility is required it may be necessary, when employing the acid cupola, to combine the magnesium procedure with a small addition of cerium. Annealing will be necessary. Some of the mechanical and engineering properties of the iron are described, with a theoretical consideration in respect of the formation of spheroidal graphite structures. (42 references).—P. M. C.

Modern Aluminium and Stainless Steel Foundry. (*Mech. Handling*, 1953, **40**, Sept., 416–418). Details are given of a new aluminium, copper and stainless steel foundry built for A.P.V. Paramount Ltd. at Crawley, Sussex. The foundry is completely modern in design and laid out for maximum flexibility of working.—D. H.

Comparison between Sintered and ' Lost Wax ' Process Steels from the Point of view of Tolerances and Utilizability. R. Baldini and V. di Sambuy. (*Ingegneria Meccanica*, 1953, **2**, Apr., 27–30 ; May, 5–8). [In Italian]. The authors describe the sintering and lost-wax processes for steel castings as used in Italy. In the sintering process, suitable metal powders are compressed at pressures between 4 and 8 tons/sq.cm. in appropriately designed dies. These are heated to weld together the metal powder particles. The well known lost-wax process is also described. Details are given of tolerances, shapes which can be cast, surface quality, size limitations of castings, practical applications and production costs. The authors compare the advantages and limitations of both methods.—M. D. J. B.

Stresses in Steel Castings, Cracks and Methods for Their Prevention. G. Kniaginin. (*Przeglad Odlewnictwa*, 1953, **3**, (8), 229–232). [In Polish]. Stresses in steel castings and their causes are discussed. The appearance of hot and cold cracks and methods of elimination are outlined.—V. G.

A Study of Steel Casting Technique. F. Brunn and H. Schiegner. (*Met. u. Giesserei Techn.*, 1953, **3**, June, 223–225). The causes of defects in the finished casting and shortcomings in normal processes are described ; remedies are suggested. A theory is put forward to account for the effect of graphite upon mould materials.—L. J. L.

Dimensional Tolerances for Steel Castings. C. H. Briggs. (*Mech. Eng.*, 1953, **75**, Aug., 619–621). Some notes are given on pattern materials and construction. Steel casting dimensional tolerances are discussed.—D. H.

Reducing the Internal Scrap in a Steel Foundry by Exothermic Additions to Risers. F. Dubilzig, W. Kaiser, and H. Kühne. (*Met. u. Giesserei Techn.*, 1953, **3**, Aug., 329–333). As in the Thermit welding process, the addition to the riser will develop heat during casting owing to the oxidation of aluminium. The melt in the risers will therefore remain liquid longer, which leads to a reduction of internal scrap. Experiments are described and technical details discussed.

Tolerances and Specifications of Malleable Castings. J. H. Lansing. (*Mech. Eng.*, 1953, **75**, Aug., 621–624). Size tolerances and shrinkage allowances as applied to malleable iron castings are discussed together with some A.S.T.M. mechanical property specifications. A note is given of the application of statistical quality control to malleable iron foundries. Mention is made of low temperature impact testing of sample castings.

Recent Advances in Pattern-Making. L. Hankart. (Inst. Brit. Foundrymen : *Engineer Foundryman*, 1953, **18**, Aug., 81–95). The author describes accurate methods of making metal patterns and core boxes by taking plaster casts. Examples of the clay-plaster process are given and the use of low-shrinkage casting resins is discussed.—A. D. H.

The Design and Choice of Cross Sections for Gating Systems Fed from a Ladle Spout. K. Hess. (*Przeglad Odlewnictwa*, 1953, **3**, (7), 211–216). [In Polish]. A method of calculating narrow cross sections of gating systems is outlined. Nomograms facilitating the use of the formulæ derived are presented.—V. G.

Radio-Frequency Heating of Foundry Cores. H. E. Schoch. (*Brown Boveri Rev.*, 1952, **39**, Nov.–Dec., 421–425). The author discusses the basic principles of radio-frequency heating and describes the application of this method to the curing of cores bonded with synthetic resins.—A. D. H.

Dielectric Core Baking. (*Metal Ind.*, 1953, **83**, July 17, 45, 46). The use of synthetic resins of the urea formaldehyde and phenol formaldehyde types for core bonding in conjunction with high frequency dielectric heating is discussed. Drying time is reduced from hours to a matter of minutes with consequent reduction in costs. The process and equipment are described and illustrated.—P. M. C.

Some Experiences from Rationalizing Moulding for Large Castings. G. Åberg. (*Gjuteriet*, 1953, **43**, Aug., 137–142). [In Swedish]. The AB. Karlstads Mekaniska Verkstad reports marked success when employing moulds built up of cores instead of ramming a complete pattern. Especially suitable for large castings weighing from 200 lb. to over 10 tons, the method has many advantages over conventional dry sand moulding. Typical castings are illustrated and their manufacture briefly described. Good results were achieved when using cement-sand moulds, in which pure silica sand was mixed with 8 vol.-% cement and 5–10 vol.-% water.—G. G. K.

Shell Moulding of Stainless Steel. W. H. Dunn and R. E. Day. (*Product Eng.*, 1953, **24**, Sept., 129–133). The design of parts for shell moulding is discussed. Because of the smooth shell, increases in the length of thin sections are possible and lower pouring temperatures can be used resulting in a finer grain structure. Good tolerances with stainless alloy, such as $\pm$ 0·002 in. on a ¼-in. section can reduce the cost of finished parts. The process is very suitable for mass production and mechanization.—A. M. F.

Shell Moulding. R. Chudzikiewicz. (*Przeglad Odlewnictwa*, 1953, **3**, (8), 243–246). [In Polish]. The process of shell moulding is described.—V. G.

Investment Castings: They're Getting Bigger. J. R. Burns. (*Product Eng.*, 1953, **24**, Apr., 137–140). Examples of investment castings up to 16 in. in dia. and up to 35 lb. in weight are given. Advance in size and density is obtained by applying pressure to the metal while solidifying ; air, moisture, and combustible materials are kept away from the metal during all stages of the process, and the temperature of the metal is controlled in each phase.—A. M. F.

Centrifugal Casting in a Sand Mould. L. J. LeBlanc. (*Canad. Metals*, 1953, **16**, June, 34–36). A machine for casting soil pipe centrifugally in sand moulds has been installed by The Anthes-Imperial Co. Ltd., Ontario. An electronic control panel automatically directs 45 timed operations to produce simultaneously four 5-ft. lengths of 4-in. pipe of extreme accuracy in weight and dimension in less than 2 min. The pipe is marketed under the trade name ' Tru-Spun.'—J. C. B.

A Modernized Small Foundry. (*Canad. Metals*, 1953, **16**, June, 30–32). Otaco Ltd., Orillia, Ontario, have modernized their foundry and the plant is described. The cupola, which is a basic one, has a charging system almost completely outside the foundry building. Each ladle of molten iron is weighed and its weight automatically recorded to aid in the

metallurgical control of the production of the company's ductile iron, 'Ductalloy'. There are also two twin Handy-Sandy overhead sand hopper units, a new sand recovery plant, and a Rotoclone air filter exhaust system.—J. C. B.

New Mechanised Foundry in Sydney. (*Australian Found. Trade J.*, 1953, **4,** Apr., 11, 12). **The Lister Blackstone Grey Iron Foundry.** (*Australasian Eng.*, 1953, May 7, 53–55). Brief details are given of this newly completed and fully mechanized foundry, on the occasion of its official opening. The plant is claimed to be as modern as any in the world.

Rationalizing Materials Handling in Foundries. G. Lindgren. (*Gjuteriet*, 1953, **43,** July, 119–125). [In Swedish]. Careful analysis of handling and transport methods in the Norrahammar foundry showed that handling costs in the malleable iron foundry were 82%, as compared with 18% for direct labour. After partial mechanization of certain operations in the grey iron foundry, including pneumatic arrangements and conveyor belts for sand transport, fork-lift trucks to carry ladles, and re-organizing flow routes in the trimming section, costs were found to be 72% and 28% respectively. Handling operations are analysed and new layouts shown by diagrams.—G. G. K.

Rationalization without Large Capital Investment. C. G. Söderlund. (*Gjuteriet*, 1953, **43,** June, 103–107). [In Swedish]. On an average, 97% of all foundry costs can be reduced, such reduction being obtained by efficient organization, production planning, cost control, and motion studies. The Huskvarna foundry, by changing from hand to machine moulding, reduced man hours by 51%. Tables show that the specific individual output was increased by 500% after partial mechanization, including the supply of moulding machines and re-planning of work. Full mechanization gave a total cost reduction of 44%.—G. G. K.

Sources of Error in Steel and Grey Iron Castings. H. Souresny. (*Osterr. Masch. Elektrowirtschaft*, 1953, **8,** Jan., 16–17). Defects found in steel or grey iron castings are discussed, particularly porosity and blowholes. Causes are indicated, and remedies suggested.—L. D. H.

Gas and Inherent Blowholes in the Foundry. A. Portevin. (*Rev. Mét.*, 1953, **50,** July, 445–464). Inherent blowholes (due to the state of the metal) are a result of dissolved gas, evolved by reactions such as that between ferrous oxide and carbon in steel, or of gas held in suspension. The nature of such blowholes is discussed together with the chemical, mechanical, and physical methods which may be employed to prevent their formation.—A. G.

Grinding + Mechanization Equals 2 × Cleaning Output. W. M. Fitzsimmons. (*Iron Age*, 1953, **171,** May 28, 132–134). The installation of wing frame grinders in place of hand chipping with pneumatic tools has considerably speeded up the cleaning of gray iron cylinder blocks, and manual handling has been reduced by the installation of mechanized conveyors at the International Harvester Co. Productivity has been considerably increased and cleaning is more thorough.

Better Parts Flow Trims Foundry Costs. H. Chase. (*Iron Age*, 1953, **171,** Apr., 30, 97–101). The new casting cleaning plant of Buick Motor Division, General Motor Corp. is described. The plant is highly mechanized with belt and roller conveyors, air hoists, overhead hook and chain conveyors and small cranes. Shot-blasting cabinets are used for cleaning large parts and power-driven hand tools are used for many of the lighter jobs.—A. M. F.

HEATING FURNACES AND SOAKING PITS

Instrumentation and Automatic Control of Soaking Pits. F. J. Cunningham. (*Instrument Eng.*, 1953, **1,** Oct., 72–78). The introduction of complete instrumentation and automatic control of soaking pits in steelworks is comparatively recent and mainly restricted to new installations. Many furnaces are still operated by manual control, using few instruments to aid the pit operator. The author describes some recent examples of new pits equipped with full instrumentation and automatic control.—J. C. B.

Industrial Furnace Design and Application. L. F. Spencer. (*Steel Processing*, 1953, **39,** June, 286–295; July, 345–348). The author describes at length electrical, gas-firing, and oil-firing methods of furnace heating, and the merits and demerits of each. Furnace materials (brickwork, structural steelwork) are next considered, and the types and applications of protective atmospheres are discussed. Many types of furnace are described and illustrated.—P. M. C.

New Non-Scaling Reheating Furnace. (*Metal Treatment and Drop Forg.*, 1953, **20,** June, 281, 282). A process for heating steel to temperatures up to 1250–1300° C. without oxidation by direct firing is described. Fuel gas is burnt with oxygen to the extent that only approximately half of the calorific value of the gas is used, the products of combustion being non-oxidising. These products are then burnt in an adjoining chamber with air, producing a temperature of 600–700° C. for preheating purposes. The consumption of oxygen amounts to approx. 400 cu. ft. per 1000 cu. ft. of town's or coke-oven gas.—P. M. C.

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

Isoperm—Isothermal Hardening Process to Improve Toughness and Wear Resistance of Alloy Steels. L. Dainelli and J. Tavolazzi. (*Ing. Mecc.*, 1953, **2,** Mar., 33–38). [In Italian]. A hardening process is described which, the authors claim, greatly improves the efficiency of drill bits.—M. D. J. B.

Batch Heat Treating Is Versatile. F. Crahen. (*Steel*, 1953, **133,** Sept. 28, 86–87). Flexibility is the aim achieved by the plant described for carburizing, hardening, drawing, annealing and normalizing a variety of aircraft, car, and other precision steel parts. Output is about 800,000 lb./month.

Low Temperature Treatments Improve Products and Processes. W. H. Miller. (*Iron Age*, 1953, **171,** May 28, 121–125). The cold treating of metal parts at temperatures down to −150° F. is discussed. Among the cold treating applications are the dimensional stabilization of gauge blocks to eliminate all retained austenite and the improvement of machinability and wear resistance of carburized alloys. Expansion fitting is also a clean process with no distortion, discolouration, or reduction in hardness.—A. M. F.

Heat Treatment of Tool Steels. J. G. Ritchie. (*Australasian Eng.*, 1953, Feb. 7, 73–81). The principle types of tool steels—water hardening, oil hardening, and high-speed hot work types—are reviewed. Compositions, hardening characteristics, and uses are discussed. Heat-treatment practice and equipment are considered under the headings annealing, stress relieving, hardening, tempering, and sub-zero treatments.—P. M. C.

On the Use of Alternating Current in the Electromechanical Surface Hardening of Tools and Machine Parts. L. Ya. Popilov. (*Vestnik Mashinostroeniya*, 1952, **32,** (9), 60–61). Views on the mechanism of the electromechanical surface hardening of steels are surveyed. The results obtained when using direct and alternating currents are compared. It is concluded that electrochardening equipment can be simpler and cheaper if alternating current is used.—V. G.

Gear Hardening by the Induction Method. (*Australasian Eng.*, 1953, May 7, 89, 91, 93). The three basic methods of induction hardening of gears, *i.e.*, through-hardening, contour-hardening, and flank-hardening are described, and the choice of method is discussed relative to the type and size of gear and its service requirements.—P. M. C.

Flame-Hardened Engine Parts for High Production. H. Chase. (*Mat. Methods*, 1953, **37,** Apr., 90–92). The author demonstrates the versatility of selective surface flame hardening of steel and malleable iron parts, by describing how the Buick Motor Co. treat almost all their new V-8 engine components on a production-line basis. The equipment and techniques used for hollow rocker shafts, and other parts are described and illustrated.—P. M. C.

The Importance of Heat Treatment in the Manufacture of Oil Well Sucker Rods. G. H. Dickinson. (*Metal Treating*, 1953, **4,** July–Aug., 8, 9, 34). Sucker rods are long slender shafts, $\frac{5}{8}$ in. to 1 in. in dia., 25 to 30 ft. long, which are connected in lengths, often of more than a mile, to operate pumps at the foot of the oil well. The great importance and care taken in their heat-treatment is emphasized.—P. M. C.

Automatic Heat Treatment of Crankshafts in Hagan Automatic Units. C. H. Schwerin. (*Metal Treating*, 1953, **4,** July–Aug., 2, 3). Full details and illustrations are given of an automatic heat-treating cycle for automotive crankshafts, at the rate of 300/hr. in three continuous furnace units with integral quench tanks.—P. M. C.

Better Heat Treating for Better Bearings. H. Habart. (*Iron Age*, 1953, **172,** July 16, 131–135). Heat-treatments necessary for both through hardening and case-hardening steels for ball and roller bearings are described. For through-

hardening steels martempering is becoming more widely used to reduce distortion and quench cracking. A reliable and rapid method for measuring case depth and carbon concentration by sampling, when case-hardening, is given.—A. M. F.

You Can Get Better Case Depth Measurements. A. D. Kirshenbaum and H. C. Boynton. (*Iron Age*, 1953, **171,** May 14, 138–140). Radioactive isotopes of carbon$_{14}$ and cobalt$_{60}$ have been applied to activate the carbon used in carburizing. After carburizing, the depth of case can be measured by direct contact of the steel with X-ray film. Apparatus for preparing the radioactive carbon is described and further studies such as α-iron changes and martensitic transformation are suggested as possible applications.—A. M. F.

Good Handling Speeds Heat-Treating of Small Parts. H. Chase. (*Iron Age*, 1953, **171,** May 14, 141–143). The mechanical handling at a new heat-treatment shop at the Ford Company's River Ronge plant is described. Novel handling units include : (1) A hoist and roll over for handling batches of parts in barrels ; (2) a vibrator actuated feed trough for moving parts to the furnace conveyor ; and (3) a steel conveyor for removing parts from the quench.—A. M. F.

New Processes of Case-Hardening Steels in Controlled Atmospheres. L. Bruno. (*Ing. Mecc.*, 1953, **2,** Feb., 15–22). [In Italian]. After a brief review of the conventional procedures for case-hardening and carburizing steels in controlled atmospheres, a detailed description, based on current technical literature and operating experience, is given of the 'gas self-generating furnace.' A critical examination is made of the carbonitriding process.—M. D. J. B.

Automatic Furnace Heat-Treats High-Strength Torsion Bars. R. Graham. (*Iron Age*, 1953, **171,** May 21, 129–132). A completely automatic heat-treatment plant for torsion bars up to 102 in. long by 3 in. in dia. at the Maremont Automotive Co. is described. A gas-fired, radiant tube, atmosphere-controlled hardening furnace is followed by an automatic quench and a forced convection recirculating tempering furnace. Suspension of the bars during heat-treatment prevents distortion and a tempered hardness of RC 46 to 49 is produced.—A. M. F.

Large Rotary Hearth Furnace. H. C. Knerr. (*Metal Treating*, 1953, **4,** July–Aug., 4, 5). A rotary propane-fired furnace is described, which was built for the heat-treatment of shell, aircraft, and diesel engine cylinders. The hearth diameter is 27 ft., and a manipulator is used for loading. The capacity of the furnace is between 1000 and 4000 lb./hr. depending upon size of work and heating requirements.

Annealing Type 414 Stainless Steel Forgings. A. A. Scafati. (*Mat. Methods*, 1953, **38,** July, 78, 79). The 12%-Cr martensitic type of stainless steel containing 1¼ to 2½%Ni is difficult to anneal because of the stabilizing effect of the nickel on the austenite transformation. Large ring and disc forgings up to 30 in. in dia. by 4 in. thick are being successfully annealed to the range 217–248 Brinell hardness by a two-stage operation. The forgings are heated to 1400° F., carefully soaked to remove forging strains, and then cooled to 1300° F. and held for a minimum of 14 hr. The charge is then cooled to 1240° F., held for 16 hr., furnace cooled to 1050° F., and finally rapidly cooled in air.—P. M. C.

Jet Principle Blasts Annealing Troubles. W. K. Lombard. (*Steel*, 1953, **133,** Aug. 24, 80–81). A new high-velocity burner is described which has given a production increase of over 50% when applied to a batch annealing furnace. Coils of ⅜-in. austenitic stainless steel rod are heated to 1900° F. with a uniformity within ± 5° F. Four burners are mounted at angles in each side of the furnace giving a high rate of circulation. The combined burner and combustion chamber has the following characteristics : Internal recirculation of combustion products, no secondary air, flame length 6 in., velocity 300 ft./sec., heat release 300,000 B.Th.U./hr. at 17 U.S. gall./hr. of fuel oil.—D. L. C. P.

A New Electric Car-Hearth Furnace for Stress-Relieving Grey Iron Castings and Annealing Ductile Cast Iron. G. Steiner. (*Brown Boveri Rev.*, 1952, **39,** Nov.–Dec., 425–428). An electric resistance furnace is described which can be used for stress-relieving treatments of grey iron castings at below 650° C. or the annealing of nodular iron at 840–950° C.

The Quenching Process in Fluid Hardening Media and Influencing It by the Surface Condition of the Material to be Hardened. W. Peter. (*Härterei-Technische Mitteilungen*, 1950, **5,** 65–96). The quenching power of water, oils, aqueous salt solutions, and fused salts was investigated by the silver ball method. The influence of the condition and constitution of water on its quenching power is discussed. The quenching properties of oils are described, and the use of salt solutions in producing quenching properties intermediate between those of oil and water is discussed.—L. D. H.

Heat-Treatment Processes for the Surface Hardening of Steel. A. Slattenschek. (*Härterei-Technische Mitteilungen*, 1950, **5,** 109–152). A short survey of heat-treating methods for the surface hardening of steel is first given, followed by a brief treatment of the laws governing these processes. Hardening curves showing the change in structure at various points in the cross section of the specimens are constructed, and a hydraulic model is described which illustrates the principles outlined.—L. D. H.

The Commercial Possibilities of Inductive Surface Hardening, and the Limits of its Practical Application. K. Kegel. (*Härterei-Technische Mitteilungen*, 1950, **5,** 153–173). After giving an outline of the principles of inductive surface hardening, with examples illustrating the range of the processes, the advantages and limitations from the commercial aspect are summarized.—L. D. H.

Contribution to the Theory of the Carburization of Steels. A. Schlattenschek. (*Härterei-Technische Mitteilungen*, 1950, **5,** 174–195). The commercial carburization process can be interpreted as a superimposing of two diffusion processes ; from this concept a differential equation is formulated, the solution of which is the equation of the general carburization curve. Its parameters are the characteristic values for the diffusion properties of the steel and of the carburizing action of the carburizing medium. The agreement between theoretical and practical results is illustrated by examples.—L. D. H.

The Reaction of Carburizing Agents in Carburization and Their Influence on the Surface Carbon Content of the Steel. K. Kallhardt. (*Härterei-Technische Mitteilungen*, 1950, **5,** 196–214). Carburization experiments were carried out with various agents on steel foil and turnings. In the case of substances with a carburizing level below the saturation range, the level was determined by carbonizing the test piece and ascertaining the carbon content. With the majority of carburizers, with a level above saturation point, a rough comparison of carburizing power was obtained from the velocity of carbide formation, or from the CO basis in the use of solid agents. Carburization curves are presented for various steels and carburizers.—L. D. H.

The Appropriate Hardening Treatment for Case-hardened Structural Members. W. Stuhlmann. (*Härterei-Technische Mitteilungen*, 1950, **5,** 215–225). The various heat-treatment procedures appropriate to different types of steel products and carburizing agents are described, and their underlying principles outlined.—L. D. H.

Surface Hardened Steel Parts and Grinding Cracks. H. Staudinger. (*Härterei-Technische Mitteilungen*, 1950, **5,** 226–241). The causes of grinding cracks occurring in surface hardened steel articles are analysed and the mechanism of formation is discussed in detail. Unsuitable grinding technique rather than chemical composition is a primary cause of failures. Suitable grinding procedures are recommended.

The Hardening and Refining of Cast Iron and Malleable Cast Iron. F. Roll. (*Härterei-Technische Mitteilungen*, 1950, **5,** 242–258). The hardening and annealing of cast iron and of white-heart and black-heart malleable cast iron are briefly described, from the point of view of the effect of the heat-treatments on physical properties. Flame and induction hardening of these materials are also discussed.—L. D. H.

Statistics and Heat Treatment E. Rossow. (*Härterei-Technische Mitteilungen*, 1950, **5,** 259–283). The author first outlines the advantages of statistical quality control, and explains the techniques best suited to a plant not employing statistical experts. Examples of the applications of statistical methods to heat-treatment operations are given.—L. D. H.

Gas Nitrided 4140 Case is Tougher. J. G. Morrison. (*Iron Age*, 1953, **172,** July 9, 129–134). Precision machine parts made from chromium-molybdenum steel have been successfully gas-nitrided by the Landis Machine Co. This has reduced the production costs for those parts in which aluminium-bearing steels would be too brittle or costly. There is slightly greater growth than with other nitriding steels, but a highly wear-resistant tough case is developed and there is less chance of grinding cracks. Tables showing the relation-

ships between surface hardness and impact values for various conditions are included.—A. M. F.

Electric Heat-Treatment Installations. E. Treichler. (*Brown Boveri Rev.*, 1952, **39,** Nov.–Dec., 442–451). The author briefly describes hardening, case-hardening, and nitriding of tool and other steels in electric furnaces. Reference is made to the layout of such installations.—A. D. H.

FORGING, STAMPING, DRAWING, AND PRESSING

German Experience Aids U.S. Artillery Shell Forgers. W. Trinks. (*Steel Processing*, 1953, **39,** July, 340–344). In 1944 the Verein Deutscher Eisenhüttenleute circulated a report to shell-forging works, dealing with faults in hot-forged shells caused by defective steel or by incorrect forging methods. Extracts from this report, together with many photographs of faults, are given.—P. M. C.

Forging Capacity of Hammers. K. Lange. (*Metal Treatment and Drop Forg.*, 1953, **20,** July, 310–312). German work into estimating the force of the blow of drop hammers by using lead cylinders, is described. Using lead of uniform purity and keeping all experimental conditions constant, it is concluded that the useful work done by the falling tup is given by the expression $3{\cdot}95V(-\delta_h)^{1{\cdot}37}$ where V = cylinder volume, and $-\delta_h$ = logarithmic strain.—P. M. C.

Planning Reduces Forging Costs. F. Emanuel. (*Metal Treatment and Drop Forg.*, 1953, **20,** June, 253–255, 252). Certain drop-forging methods are described which allow saving on material, by using metal which is ordinarily wasted, thus reducing costs. The production of forgings in multiples, combination forging of two dissimilar parts from one blank, and the use for smaller items of the pierced slugs obtained from ring forgings are discussed as typical methods.—P. M. C.

Large Pressure Vessel Forgings. A. A. Munro. (*Times Rev. Indust.*, 1953, **7,** Nov., 28–33). A detailed and illustrated account is given of developments since 1918 in the production of heavy steel forgings for pressure vessels. The difficulties overcome in casting 120-ton ingots, trepanning the bore, and in increasing diameter and length by forging are explained. The welded boiler drum and a German method of strip winding high-pressure forgings are also referred to.

German Research in Drop Forging. O. Kienzle. (*Metal Treatment and Drop Forg.*, 1953, **20,** Aug., 316–368). The author reviews the work in progress in the Drop Forging Research Department of the Technical University of Hanover. Equipment includes a drop hammer with 700-lb. tup, a pneumatic hammer with 200-lb. tup, a 14,000 ft. lb. friction screw press, a 250-ton hydraulic press, a 50-ton trimming press, and gas, electric and H.F. induction heating units. A brief outline is given of the field of work, which covers stress measurements on drop hammer friction belts and methods of belt fixing, tup velocities and work required for particular forging operations, the effects of mismatching of dies, die wear, and die sinking.—P. M. C.

Induction Forging Can Be Versatile. F. T. Chesnut. (*Steel*, 1953, **133,** Sept. 7, 112–113). One of the first all-induction-heated forges applied to peacetime production of car components is successfully turning out gears, steering arms, brake shafts and over 100 other parts for the Massey-Harris Co. The three 500-kW. induction units can each hold up to seven of the interchangeable heaters to suit the job. The work can be delivered automatically to the forge operator. Cost of heating is somewhat higher than using fuel, but overall economy favours induction-heating—D. L. C. P.

Cold Shaping of Steel by Compression Reaches Commercial Applications. A. H. Allen. (*Metal Progress*, 1953, **64,** Aug., 65–68). A short account of the cold extrusion of steel as carried out by the Mullins Manufacturing Corp., Salem, Ohio, is given. Ordinary mild steel is used. Accurate design of the punch and die, phosphate treatment of the steel surface to retain a lubricant, and careful control of the pressure in the different operations are the three basic requirements.—B. G. B.

New Forming Process Reduces Vibration and Tooling Costs. T. A. Dickinson. (*Steel Processing*, 1953, **39,** June, 283–285, 302). The 'Formall' process is described, in which a drop hammer is used to produce deep draws in sheet metal which would require much more powerful and expensive equipment if hydraulic pressure were used. The hammer is modified by fitting an upper (falling) forming unit, consisting essentially of rubber pressure pads, and a lower hydraulic die retainer unit with a shock absorber. Hardwood and phenolic plastic dies have been used successfully, and $\frac{3}{8}$ in. thick steel sheet has been formed with only 30 lb. of dead weight.—P. M. C.

Radial Forming by Ryan. (*Steel Processing*, 1953, **39,** July, 324—326). An ingenious new expansion forming technique for fabricating aluminium and stainless steel sheet-metal parts has been developed by Ryan Aeronautical Co. The forming is effected by a huge expanding mandrel operated by hydraulic fluid at 5000 lb./sq. in. Radial forces of 4800 tons can be exerted. The technique is used to form aircraft fuel tanks, jet engine parts, and all kinds of large contoured closed sections to exact dimensions.—P. M. C.

Progress in Metal-Forming Equipment. W. H. S. Freeman. (*Welding and Metal Fab.*, 1953, **21,** Sept., 318–323).

New Setups Speed Production of Stampings. H. Chase. (*Iron Age*, 1953, **172,** Aug. 13, 142–144). A new installation for the quantity production of impellor and turbine blades for torque converters is described. Six-stage progressive dies are used in a 75-ton press operating at 70 strokes/min. For uniformity, restrike presses are used. These are equipped with automatic pickup and loading devices fed from a hand loaded indexing dial. Plastic blocks are used on the dial to locate the parts.—A. M. F.

Press-Forming Process. (*Metal Ind.*, 1953, **83,** Aug. 28, 169, 170). The Hildraw process developed two years ago by Consolidated Vultee Aircraft Corp. is described. It is briefly a press-forming technique which permits the use of a rubber pressure pad as a female die and an inexpensive form block as a punch. A blank is drawn by moving the punch upward so as to displace the rubber die, after which the displaced rubber envelops the punch and blank, forcing the pressure pad and blank downward against hydraulic pressure. Rubber pressures range from 4000 to 10,000 lb./sq. in., the higher pressures being required for stainless steels. Presses from 1000 to 7000 tons capacity are being used.—P. M. C.

Tool Steels for Press Tools and Shear Blades in Sheet-Metal Fabrication. R. S. Watts. (*Sheet Metal Ind.*, 1953, **30,** Apr., 269–276, 280 ; May, 375–384 ; July, 553–565). A representative selection of tool steel compositions is given, and their properties and selection for particular uses is discussed. Heat-treatment is next considered, and furnace equipment is described. Pyrometry is also dealt with, and an outline of the theory and metallurgy of tool steels is presented. Normalizing, hardening, and tempering processes are described in detail.—P. M. C.

The Use of Disc Springs in the Construction of Shearing, Punching, and Drawing Tools. K. W. Michler. (*Osterr. Masch. Elekrtowirtshaft*, 1953, **8,** Feb., 38–39). The advantages of disc springs over coil springs in the manufacture of certain types of tool are pointed out, and illustrated by characteristic curves.—L. D. H.

How to Increase the Life of Cutting Dies. F. Strasser. (*Iron Age*, 1953, **171,** May 14, 144–146). Factors causing wear of punching dies are discussed. Wear depends not only on the type of material cut but the design, material heat-treatment, and the setting and lubrication of the die components. For long runs cemented carbide dies should be considered as, although the cost is high, the life is better. In die construction care must be taken to ensure alignment, correct penetration and accurate location of the tool shank.

Methods and Machines for the Testing of Materials. IX—Drawing Tests. M. C. Boulet. (*Usine Nouvelle*, 1953, Apr. 30, 31–32). All practical tests of drawability are considered to be empirical, while the apparatus required is often expensive. An equation has been derived relating drawability to a workhardening factor calculated from the normal tensile test. For deep drawing, additional tests are needed as grain texture and surface hardness become important. The required information can be obtained from a Rockwell hardness test and micrographic examination.—A. G.

Die Design for Metal Drawing. C. R. Cory. (Amer. Soc. Tool Engineers : *Steel Processing*, 1953, **39,** June, 269–272). The author discusses panel drawing operations with reference to wrinkling and tearing. The control of pressure on the blank holding plate is dealt with, and die designs for automobile door panels and roofs are illustrated and described.

Coated Steels Can Cut Breakage on Drawn Parts. N. E. Hays. (*Iron Age*, 1953, **172,** Aug. 13, 135–137). The advantages of zinc and zinc phosphate coating on steel for drawing operations are discussed. Wedge tests and production tests on a deep drawing enamelling iron have shown that there is

far less tendency for breakage in the dies. The function of the coating in the draw appeared to be to introduce the lubricant more deeply into the die, which eliminated points of strain.—A. M. F.

Factors Influencing Cold Extrusion of Steel. H. J. Pessl and H. H. Hauttmann. (*Metal Progress*, 1953, **64**, Aug., 69–73). The results of cold extrusion tests which have been carried out on 1½ and 4 in. bars of 11 different types of steel are given. The effect of die angle on toughness, hardness and extrusion pressure for these steels and the effect of reduction of area on notch toughness are also shown.—B. G. B.

Tools, Lubricants, and Steels for Cold Extrusion. T. Bishop. (*Metal Progress*, 1953, **64**, Aug., 74–77). The selection, use, and heat-treatment of 16 different steels suitable for the fabrication of dies are discussed. The properties of 11 steels suitable for cold extrusion are tabulated.—B. G. B.

Some Observations on the Metallurgy of the Deep Drawing of Metals. D. V. Wilson. (*Sheet Metal Ind.*, 1953, **30**, Aug., 621–629, 640). The testing of drawing capacity is discussed, and details are given of a form of test developed by the author and applicable to thick strip. A stepped test-piece of circular cross-section is machined from the strip, the eight steps having diameters increasing progressively from 0·275 to 0·506 in. The increase in cross-sectional area is 10% at each step. The lubricated test piece is drawn through a round die (0·275 in. dia.) in a tensile testing machine. Drawing is continued until the material held in the grips (0·275 in. dia.) fractures. The grips are then advanced to the first section which has received 10% reduction and drawing is continued until this section in turn fractures. The procedure is continued until fracture occurs in or near the die. Readings are taken of tensile, *i.e.*, drawing stress, and percentage elongation on each section. Typical results of such tests, and attempts at correlation with drawability of the sheet are discussed. The effect of the die angle is examined.—P. M. C.

On the Profile of Tube Drawing Dies. I. Measuring Method of Die-Profile. S. Yazawa and K. Sato. (*J. Mech. Lab.*, 1953, **7**, Mar., 63–66). [In Japanese]. Work on die profilometers is reviewed. An instrument has been made up, using the principles of the Yamamoto roughness tester. Results are presented for die profiles and wear in tungsten carbide and steel dies.—K. E. J.

A Method of Measuring the Surface Temperature of a Wire During Drawing. S. Z. M. Kopczynski and J. S. Hoggart. (*Australasian Eng.*, 1953, Apr. 7, 75–78). Methods of measuring the surface temperature of a wire during drawing are discussed critically and conditions which such a method must satisfy are proposed. A method using a composite steel-constantan die is described which satisfies these conditions. An annulus of constantan 0·003 in. thick is clamped between two halves of a carbon steel die, the junction being arranged to coincide with the threat of the composite die. The method of calibration is described.—P. M. C.

ROLLING-MILL PRACTICE

Electrical Equipment for the Steel Company of Wales. P. C. Dannatt. (*Metro. Vick. Gaz.*, 1953, **25**, July, 3–22). The mill motors and control gear at the Abbey and Trostre works of the Steel Company of Wales are described in detail.

Electronic and Magnetic Methods of Controlling Continuous and Reversing Hot Rolling Mills. G. Lemcke. (*Stahl u. Eisen*, 1953, **73**, Aug. 27, 1156–1161). Examples are given to show how various control methods can comply with the requirements of rolling mills. The reliability of electronic control has been proved ; it competes with magnetic amplifier regulation. Both serve for the control of D.C. motors through rectifiers, the use of which has decisively influenced German rolling-mill technique.—J. P.

Modern Drives and Controls of Heavy Reversing and Cold Rolling Mill Trains. H. Nüsslin. (*Stahl u. Eisen*, 1953, **73**, Aug. 27, 1148–1155). Recent improvements in rolling-mill drives are reported. A study of centrifugal forces in reversing and accelerating drives has led to the development of new types of motor construction. Twin drive is being more widely employed. The whole machine bay, including the cellar, is being included in the ventilation system and, by introducing circulating air cooling, the old problem of the contamination of electrical drives is being solved. Mechanical and magnetic amplifiers make semi-automatic control possible, and this has lightened the duties of the operator and increased operational safety. Control in cold rolling mills is discussed and some results with a modern four-high mill are reported ; a considerable increase in output has been achieved merely by reconstructing the electrical controls.—J. P.

Examples of Electronic Control and Regulating Devices for Drives in German Rolling Mills. H. Opitz. (*Stahl u. Eisen*, 1953, **73**, Aug. 27, 1141–1148). Examples of modern rolling-mill drives with electronic control and regulating devices are described. These comprise a cogging-mill reversing drive, a continuous strip mill train with single drives, roll-setting, and other D.C. auxiliary drives, all with rectifier control, and follower or remote controls of high accuracy for use in various parts of the mill. The electrical circuits are illustrated and their performance fully described.—J. P.

Pressure, Spread, and Stretching in Hot Rolling. Z. Wusatowski. (*Met. u. Giesserei Techn.*, 1953, **3**, June, 213–220). From experimental results and mathematical calculations a formula has been derived for the coefficient of spread, which also permits determination of the coefficient of stretching. The maximum error in applying the formula was 4·59%. In normal conditions effects of temperature, rolling speeds between 0·4 and 7 m./sec., roll quality, roll surface, and type of steel upon the spread formula can be neglected.—L. J. L.

A Study of Primary Mills. W. Craig. (*Anvil*, 1953, **3**, Autumn, 4–9). The author gives a comprehensive review of the manufacture of primary rolling mills by the Davy United Group. A detailed analysis is made of the components of slabbing and blooming mills, and the changes which have been effected to improve their efficiency are noted.—J. C. B.

Knurling Cogging Rolls. A. H. Norris. (*J. Iron Steel Inst.*, 1954, **176**, Feb., 157–158). [This issue].

Calculation of Job Times by Using Prime Records of Operations in Rolling Mills. H. Gobiet and K. Schnick. (*Stahl u. Eisen*, 1953, **73**, Sept. 10, 1226–1231). A method is described by which job times can be evaluated without resort to time and motion studies. The procedure applies only to existing jobs, not to new ones.—J. P.

Historical Evolution of Blooming and Slabbing Mills. G. Grenier. (*Echo Mines*, 1952, Oct., 649–651 ; Nov., 721–725 ; Dec., 785–788 ; 1953, Jan., 27–28 ; Feb., 99–101 ; Mar., 167–168 ; Apr., 239–241 ; May, 307–310 ; June, 391–392). This is a comprehensive account of the development of the rolling of blooms and slabs. It is arranged in the following sections : (A) General history ; (B) auxiliary machinery ; (C) blooming and slabbing mills in 1900 ; (D) history of mill stands, motors, and controls ; (E) ingots for rolling ; (F) roll pressure and metal deformation ; (G) flaws and other defects ; (H) blooming mills for carbon steel ingots ; (K) blooming mills for special steel ingots ; (L) composition of rolls and pass design ; (M) entering the ingot ; (N) effect of height of the upper roll ; (O) conclusions ; and (P) examples of installations.—R. A. R.

Rolling Billets, with Restricted Steel Ingot Production, into a Wide Range of Products. W. Kruse. (*Stahl u. Eisen*, 1953, **73**, Aug. 13, 1101–1107). In 1951, approximately 60% of German ingot production was converted to semi-finished products. The conditions in the most important German semi-finished products works, including two dismantled plants, are described and the differences between the structures of American and German markets and industries are studied. These differences must lead to different solutions of technical problems. Fully continuous mills need large outputs and a narrow range of products. German conditions necessitate a middle way, in which semi-continuous mills are to be preferred. The lay-out of a mill with limited production and a wide range of products is described.—J. P.

Rolling Mill Yield. L. R. Silliman. (*Iron Steel Eng.*, 1953, **30**, July, 103–109). The author describes some experiences and studies of fundamental practices affecting billet and bar yields. Yield is increased as ingot delivery time is increased and is decreased as time in soaking pits is increased. Yield variation with soaking pit practice is examined.—M. D. J. B.

The Construction of the Rolling Mills at Linz. H. Weitzer. (*Osterr. Masch. Elektrowirtschaft*, 1953, **8**, Jan., 6–8). The layout and production figures of the sheet rolling plant at the Linz Works of Vöest are presented and discussed.—L. D. H.

The Milling of Rail Ends and Its Effect on Productivity. Sir Charles F. Goodeve. (*Engineering*, 1953, **176**, Sept. 4, 299). The author examines the relative merits of hot-sawn rails as produced in the U.S.A., and the British practice of milling all rails. It is concluded that the British practice gives a better product without loss of productivity.—M. D. J. B.

The Temper Mill Drives at the Trostre Tinplate Works. (*Engineer*, 1953, **196,** Aug. 28, 265–268). This article describes the two-stand temper mills at the Trostre tinplate works of the Steel Company of Wales Ltd., which operates at speeds of up to 4000 ft./min. In the electrical equipment a cascade control exciter system is applied in conjunction with Ward-Leonard control for maintaining the correct strip tension.

New Continuous Rolling-Mill Plant Started in Holland. (*Sheet Metal Ind.*, 1953, **30,** Aug., 681–685). A brief description with illustrations is given of the Dutch State Steelworks, Breedband N.V., at Ymuiden. The four-stand hot-strip mill is semi-continuous, and capable of producing 0·07 in. gauge by 36 in. wide strip. The four-stand tandem cold mill will roll tinplate down to 0·008 in. gauge ; the roll bodies are 56 in. long.—P. M. C.

High Speed Mills and Their Application to Ferrous and Non-Ferrous Rolling. G. Perrault, jun. (*Iron Steel Eng.*, 1953, **30,** July, 83–89). The author describes the refinements which have to be brought into operation with very high speed mills running at speeds up to 7000 ft./min. ; these, in addition to the more obvious machining refinements, top quality bearings, dynamic balancing of rotating parts involve careful designing of materials handling equipment and installations.—M. D. J. B.

Pittsburgh Steel Starts New 66-in. Hot Strip Mill. (*Iron Steel Eng.*, 1953, **30,** Aug., 118–124). A description is given of a new 66-in. semi-continuous hot strip mill at the Allenport works of Pittsburgh Steel Co. Ultimately the mill will have an output of 600,000 tons per year. Sheets 20 to 60 in. wide and 0·065 to 0·188 in. thick can be rolled.—M. D. J. B.

Precision Rolled Thin Strip Offers: 1. Close Tolerances; 2. Superior Surface Qualities; 3. Controlled Grain Size and Hardness. A. I. Nussbaum. (*Mat. Methods*, 1953, **38,** July, 82–84). A brief account is presented of the types and gauges of ultrathin strip now commercially available through the use of the Sendzimer cluster mill. Gauges down to 0·00015 in. can be produced, to close tolerances and with superior surface finish, and controlled grain size and hardness. Some important applications of such strip, made from a variety of metals and alloys, are instrument membranes, thickness gauges, camera shutters, diaphragms and bellows, springs, and sliding contacts.—P. M. C.

Replacements for Palm Oil in Cold Rolling Steel. R. C. Williams. (*Iron Steel Eng.*, 1953, **30,** Aug., 65–66). The author examines the possibilities of replacing palm oil for cold rolling processes by other lubricants of domestic origin. The requirements, it is suggested, can be met by an oil supplied by the American Oil Chemists' Society.—M. D. J. B.

MACHINERY FOR IRON AND STEEL PLANT

Crane Gantry Girders for Steelworks. J. S. Terrington and J. M. Hawkes. (*Structural Eng.*, 1953, **31,** Oct., 268–285). An attempt is made to show how improved design and resulting economy can be achieved in crane gantry girders by closer adherence to fundamental theory and to the best practice. A single-web plategirder is shown to be preferable to other types and a formula for a girder of least weight is given. For the flanges, a new development of the strain energy equation is put forward in graphical form.—J. C. B.

New Safety Hook Block Will Reduce Crane Accidents. (*Iron Steel Eng.*, 1953, **30,** July, 116–118). This device consists of a spring system introduced between the hook block and the lifting beam. The springs are designed with sufficient capacity to absorb the kinetic energy of all rotating and moving parts.—M. D. J. B.

The Scraper Conveyor of the Vereinigte Österreichische Eisen- und Stahlwerke A.G. J. Zeman. (*Österr. Masch. Elektrowirtschaft*, 1953, **8,** Aug., 252–254). Details of design and performance of scraper chain conveyors manufactured by Vöest are given. Photographs and diagrams are reproduced.—L. D. H.

Internal Transport as a Technical and Economic Problem for Iron and Steel Works. K. Rosenbaum. (*Stahl u. Eisen*, 1953, **73,** July 30, 1046–1053). Works transport plays an important part in the steelworks ; its cost represents a high proportion of production overheads. The application of modern supervision methods have increased efficiency in some works by 40% ; the frequency of movement of wagons could be raised sufficiently to avoid the otherwise necessary extension of the wagon yard. The characteristics of steam, Diesel, and electric locomotives are compared. Apart from special cases, the electric locomotive is superior to the other types. Radio-communication has proved best for remote control of railway operation.—J. P.

Increasing the Efficiency of Railway Transport in Iron and Steel Works. H. Bergrath. (*Stahl u. Eisen*, 1953, **73,** July 30, 1053–1057). Modern tendencies to increase output rates puts a premium on smooth flow of materials and thus makes a study of traffic problems essential. The fundamentals of internal transport, the significance of human collaboration, and factors influencing costs and efficiency are discussed. The characteristics of various communication systems for railway operation are compared.—J. P.

Synchronous Motor Torque Requirements for Steel Mill Applications. W. A. Thomas. (*Iron Steel Eng.*, 1953, **30,** Aug., 73–81). This paper reviews the methods of obtaining synchronous motor torques for various applications, to help users to anticipate their needs, and designers to satisfy requirements. The author suggests that load requirement is adequately defined by the three cardinal values of starting torque, pull-in torque, and pull-out torque. Tables are given indicating the Amer. Inst. of Electrical Engineers and the National Electric Manufacturers Association's motor specifications for various duties.—M. D. J. B.

Dust-Free Ventilation of Electrical Machines with Particular Reference to Steelworks' Plant. W. Schemmerling. (*Stahl u. Eisen*, 1953, **73,** Aug. 27, 1162–1168). The dust content of the air in workshops is much higher than is generally assumed. Protection of electrical machines and apparatus from dust is therefore of urgent importance. Efficient operation of heavy duty motors requires that they be well cooled. Apart from water cooling of large turbines, filter or circulating coolers, usually connected with separate ventilation by radial or axial fans, offer a satisfactory solution. Modern filtering techniques also offer economic methods for cooling motors. When a large number of motors is to be cooled, a central compressed air plant with rotating filters is often to be preferred. For high voltage switch rooms, erected in dusty places, only high pressure plant can be considered.—J. P.

Trends in Electrical Equipment for Materials Handling in the Steel Industry. S. Rifkin. (*Iron Steel Eng.*, 1953, **30,** Aug., 84–90). Reference is made to fork lift trucks, conveyor belts, cranes, coal and ore bridges, unloaders and blast-furnace skip hoists. The principle trends are : Use of standardized components (such as motor controls), adjustable-voltage speed control, greater safety, creeping speed control, and adjustable-voltage drives.—M. D. J. B.

Cast Meehanite Rolls Used for Heavy-Duty Bending. M. R. Nelson. (*Iron Age*, 1953, **171,** May 28, 135–137). The design and construction of two rolls for bending plates up to 2½ in. thick is described. The rolls are 10 ft. 2 in. between frame sections and weigh 247,000 lb. Gearing for the pull down of the tap roll is also made from heat-treated Meehanite castings. This material has saved the high cost of alloy steel forgings originally thought necessary for the great stresses involved.

The Magnetic Amplifier in Control Circuits. R. W. Moore. (*Iron Steel Eng.*, 1953, **30,** July, 67–73). The author reviews magnetic amplifiers used in the iron and steel industry and discusses their relative merits. Amongst the types examined are the half wave self-saturating and the parallel self-saturating magnetic amplifier, used as regulators to maintain constant strip tension, a current limit device, and voltage regulators.—M. D. J. B.

LUBRICANTS AND LUBRICATION

The Multi-Purpose Grease Approach to Lubrication. G. H. Link. (*Product. Eng.*, 1953, **24,** July, 174–177). The properties of different types of grease are described and compared with a lithium type multi-purpose grease. This is a product combining lithium and a 12-hydroxy stearic acid. These greases have the best available low temperature characteristics and are good for high temperature applications. Certain types have the highest known mechanical stability.—A. M. F.

How Colloidal Graphite Protects Bearing Surfaces. H. Warburton. (*Iron Age*, 1953, **172,** July 9, 135–136). The advantages of colloidal graphite as a lubricant are discussed. It will function as a lubricant well above the boiling point of oil and can withstand temperatures of 600° C. in a normal oxidising atmosphere. It is readily absorbed by most metals and is applicable in a dry solid lubricant as well as in water, oil or alcohol.—A. M. F.

The Lubrication of Ball and Roller Bearings. Parts II, III, IV and V. (*Sci. Lubrication*, 1952, **4,** Nov., 12–16 ; 1953, **5,**

Feb., 17–23 ; Mar., 17–24, Aug. 10–15). In most cases, oil in some form is the best lubricant for anti-friction bearings ; oil spray or mist is ideal. Manufacturers' recommendations are reviewed. A chart shows the most suitable viscosity for various sizes and speeds of bearings. Methods of oil lubrication are described. Reasons of cost, configuration or temperature may lead to the adoption of greases : properties and advantages are discussed for lime, soda, aluminium, barium, lithium and graphite based greases.—K. E. J.

WELDING AND FLAME-CUTTING

Recent Developments in the Fusion Welding of Iron and Steel. J. W. Shedden and W. I. Pumphrey. (*Metallurgia*, 1953, **48,** Aug., 78–84). Recent advances in the fusion welding of iron and steel are described. The increasing use of low-hydrogen electrodes for the arc welding of low-alloy steels is considered. The use of argon containing a small amount of oxygen has made possible the welding of mild steel by the inert-gas-shielded metal-arc process. The development of ferrite-controlled metal-arc electrodes for the welding of stainless steels is also described.—B. G. B.

Developments in Metallic Arc Welding of Corrosion- and Heat-Resisting Alloys. M. C. T. Bystram. (*Welding Metal Fab.*, 1953, **21,** Sept., 325–326). Methods of preventing intergranular corrosion in the welding of stainless steels are pointed out and the possibilities of mixed welding, *i.e.*, making most of the weld with carbon steel electrodes and finishing with stainless or other high alloy steel, are shown. The difficulties of making a joint between a low alloy air-hardening steel and a fully austenitic stainless steel are explained with recommendations for overcoming them.—R. A. R.

Conference on Welded Structures, 1953. (*Inst. Civil Engineers*, 1953). This Conference, organized by n Joint Committee of the Ministry of Works, Institution of Civil Engineers and Institution of Structural Engineers, was held in London on Nov. 23–26, 1953, to review present knowledge and outstanding problems in the design, fabrication, and erection of welded steel structures. The papers presented were :

General Review of Welding in Shipbuilding. F. C. Cocks.

Problems of Fabrication and Erection in All-Welded Multi-Storey Framed Buildings. D. C. C. Dixon.

The Design of Welded Steel Frames for Multi-Storey Buildings. W. A. Mitchell.

Dock Gates and Caissons. Part I—Welded Dock Gates. F. J. Walker.

Oil and Gas Storage Vessels. M. Noone.

Design of Single-Storey Framed Buildings and Portal Frames. W. S. Atkins.

Fabrication and Erection of Single-Storey Framed Buildings and Portal Frames. R. W. Schofield.

Design, Fabrication and Erection of Braced Roof Structures. S. M. Reisser.

Plate-Girder Bridges. G. Roberts and O. A. Kerensky.

Design, Fabrication and Erection of Trussed Girder Bridges. E. M. Lewis.

Welded Tubular Structures. T. Bedford.

Recommendations Prepared by Committee No. 9 (Weldability) of the Institut International de la Soudure. (*Rev. Soudure*, 1953, **9,** (2), 83–87). These recommendations were elaborated to define the minimum requirements of carbon and low-alloy manganese steels to ensure adequate weldability by the electric arc process. Quality requirements for thicknesses, mechanical properties, and chemical compositions are specified.—M. D. J. B.

A Proposed Classification for Weldability Tests. H. Granjon. (*Rev. Soudure*, 1953, **9,** (2), Supplement 1–13). The author stresses the need for and proposes a classification of weldability tests.—M. D. J. B.

Fillerarc Welding Process. R. W. Tuthill. (*Welding J.*, 1953, **32,** Aug., 703–707). Fillerarc is the name of the General Electric Company's new equipment for consumable electrode gas-shielded welding. It consists of a gun, a wire feeding unit, and a specially designed welding generator. A deta.led description is given.—U. E.

Effects of Electrodes and Heat Treatments on Mild Steel Weldments. E. F. Nippes and A. Lesnewich. (*Welding J.*, 1953, **32,** Aug., 353s–374s). The ductility of welds made with E6010 and E6016 electrodes on three plain carbon silicon-killed steels 1 in. thick was studied using the Kinzel test. In the as-welded condition E6016 was superior to E6010 only with the steel containing 0·25% C. For all other conditions of preheat and post-heat investigated there was no difference in the transition temperature between deposits from E6010 and E6016. Stress-relief for 1 hr. at 1150° F. was found to give best results.—U. E.

Stranded Electrode Materials for Aircomatic Welding. H. Robinson and H. C. Cook. (*Welding J.*, 1953, **32,** Aug., 692–702). The fabrication and use of stranded electrodes are discussed. Several alloys which cannot be drawn into continuous lengths of wire for Aircomatic application can be supplied in composite form. Several aluminium–bronze compositions and a lead-tin bronze are available. The composite electrode consists of strands of several pure metal wires in the proper proportions for the desired alloy. Tests have shown that the wire elements mix thoroughly and alloy together in the arc and deposit. The properties of the deposit compare favourably with those obtained from solid alloy wires.—U. E.

Automatic Welding Improves Stainless Tube Output. (*Steel*, 1953, **133,** Sept. 21, 190–191). An increase in automatic inert-gas non-consumable electrode welding of stainless steel tube is reported, and an installation is described. Speeds up to 96 in./min. are obtainable with thin sections.—D. L. C. P.

Welding the Stainless Steels. H. Thielsch. (*Mat. Methods*, 1953, **37,** May, 115–130). The compositions and characteristics of the three types of stainless steel are briefly outlined, and full details are given of the types, compositions, and applications of welding electrodes suitable for welding them. Procedures and precautions necessary for the satisfactory welding of austenitic, martensitic, and ferritic grades of stainless steel are discussed.—P. M. C.

The " Eutectic " Low Temperature Welding Process. (*Australian Found. Trade J.*, 1953, **4,** July, 19–22). A new series of welding rods has been developed which, because of the affinity of the rod alloy for the parent metal, allows bonding at temperatures well below the melting points of either. A complete range of 46 alloys is available suitable for welding many different metals.—P. M. C.

Welded Hollow Turbine Rotors of Turbine Ship Andrea Doria. B. Conti-Barbaran. (*Ingegneria Meccanica*, 1953, **2,** July, 17–23 ; Aug. 17–22). [In Italian]. The author describes the hollow welded turbine rotors for a ship and considers solutions to this problem which might have been adopted. Information is given on the type of steel used and on the welding procedure adopted ; fabricating methods, handling, and inspection of the rotors are also described.

Low Temperature Stress Relieving of Mild Steel Welded Structures. L. E. Benson and S. J. Watson. (*Trans. Inst. Weld.*, 1953, **16,** Aug., 90–95). An investigation was carried out on the stress relieving of mild steel welded structures. The reduction in average stress of a welded plate by annealing at several different temperatures was determined. For normal conditions in practice the efficiency of stress relief at 200–300° C. is about 10%. The degree of initial plastic strain and associated strain hardening effects have a marked effect on the residual stress value after stress relieving at 200–400° C., but this is not the case when stress relieving is carried out at 600° C. Markedly different characteristics were found when comparing weld metal with wrought mild steel, high stresses may still exist in welded joints if these are only stress relieved at low temperatures.—U. E.

The Welding of Certain Heat-Resisting Steels. J. A. McWilliam. (*Trans. Inst. Weld.*, 1953, **16,** Aug., 96–99, 110). The metal arc welding of a steel of low-carbon 13% Cr steel is compared with a similar steel containing 0·16% aluminium. Cracks occurred in the 13% Cr steel when arc welded with electrodes of similar composition even when preheat and postheat were carried out. With 18/8 and 25/17 Cr/Ni electrodes no cracks were found. The addition of 0·16% Al appeared to have little effect in lessening hardening.—U. E.

Welding of Torsteel. S. Soretz. (*Schweisstechn.*, 1953, **7,** July–Aug., 77–90). The flash butt welding process is discussed. The method is applied to the welding of heat-treated steel used for reinforcing concrete, *e.g.*, Torsteel. After being flash butt welded the steel is heated again and slightly twisted which improves the physical properties of the joint. The same treatment is successfully applied to fusion welded joints.—U. E.

High-Speed Automatic Projection Welding. C. E. Slade. (*Welding and Metal Fab.*, 1953, **21,** Aug., 281–286). An automatic projection welding machine developed by Philips Electrical Co. is described. The machine incorporates an

air-operated toggle press for producing part of the weldment continuously from strip.—U. E.

Flash Welding and High Speed Cold Reduction of Strip Steel for Tin Plate. J. Wargo and R. C. Brunner. (*Iron Steel Eng.*, 1953, **30**, Aug., 59–64). The measures taken for reducing delay time on the high speed, cold reduction, strip mill at the Aliquippa Works of Jones and Laughlin Steel Corp. are described. These measures comprise: Flash butt welding of the hot bands by the electrical resistance method, correct positioning of the strip, carefully controlled speeds, correct composition of the strip, correct tension and proper maintenance of all machinery.—M. D. J. B.

Automatic Welding with Wire-Mesh-Wound Continuous Electrodes. C. DeRop and H. Schmidt-Bach. (*Schweissen u. Schneiden*, 1953, **5**, Aug., 315–318). The mesh-wound electrode is described. The coating fills the spaces in the wire mesh covering, the latter conducting the current from the copper discs of the automatic welder to the inside of the electrode. The coating can be either acid or alkaline.—U. E.

Conditions and Rules for Deep Penetration Welding. H. Neumann. (*Schweissen u. Schneiden*, 1953, **5**, Aug., 297–305). A review is given of the present state of deep penetration welding. Proposals are advanced for rules of testing deep penetration electrodes and rules for the welding of plates up to 22 mm. thick.—U. E.

Ultrasonic Testing of Welds. H. Krächter, J. Krautkrämer, and H. Krautkrämer. (*Schweissen u. Schneiden*, 1953, **5**, Aug., 305–314). The principles and equipment for the ultrasonic testing of welds are discussed. Welds have been tested by using vertical and inclined probes, and beam angles were determined for different plate thicknesses. X-ray and metallographic methods were applied and confirmed the defects found by the ultrasonic method.—U. E.

The Use of Structural Quality Steel in Pressure Vessels. I. E. Boberg. (*Welding J.*, 1953, **32**, Aug., 413s–416s). The use of carbon steel plates in pressure vessels and their suitability for welded construction are discussed.—U. E.

Migration of Cr and C in a Weld in Type 502. M. A. Scheil. (*Metal Progress*, 1953, **64**, Sept., 108–110). The migration of chromium and carbon in a 4–6%-Cr/Mo steel welded with a 25/20 Cr/Ni electrode after heating at 1290° F. for 10,000 hr. was investigated. The following five zones from parent metal to weld were noted : (1) Zone of well polygonized grains in parent metal ; (2) decarburized zone 0·10–0·30 in. thick; (3) narrow white zone ; (4) a darker-etching layer along junction where carbon had diffused into the weld ; and (5) columnar grained 25–20 deposit with sigma phase along grain boundaries in the bottom weld passes.—R. A. R.

An Investigation on Peening. P. L. Calamari, F. J. Crum, and G. W. Place. (*Welding J.*, 1953, **32**, Aug., 387s–402s). Modified Charpy impact tests and hardness tests were carried out on mild steel plate and weld deposits to determine the effects of peening. It was found that impact properties of plate were adversely affected by peening, especially when peening was carried out in the 500–900° F. temperature range. Tests on deposited E6010 weld metal showed that peening of the last weld layer resulted in a pronounced loss in notch toughness, but low-hydrogen E6010 welds had a much better notch toughness. The peening of plate or weld metal resulted in a steep increase in hardness.—U. E.

The Incidence of Cracking in Welding Type 347 Steels. L. K. Poole. (*Welding J.*, 1953, **32**, Aug., 403s–412s). A survey is given of the nature and occurrence of cracking difficulties encountered in welding type 347 stainless steel (C 0·08% max., Ni 9–12%, Cr 17–19% + Nb).—U. E.

Welding Metallurgy of Nodular Cast Iron. E. E. Hucke and H. Udin. (*Welding J.*, 1953, **32**, Aug., 378s–385s). The structures of two heat-affected zones and weld metal in nodular cast iron with and without pre- or post-heat are described. It was found that pre- or post-heat treatment did not render the fusion zone ductile.—U. E.

Safety Devices in Gas Plants, their Value in Welding and Cutting Operations. M. Konschak and W. Scheruhn. (*Schweissen u. Schneiden*, 1953, **5**, Sept., 347–350). Various safety devices for preventing acetylene from burning back are described in detail with drawings.—R. A. R.

Brazing by Induction. D. Warburton-Brown. (*Welding and Metal Fab.*, 1953, **21**, July 238–241 ; Aug., 291–295). The application of induction heating to industrial brazing problems is discussed. Examples are given of brazing by induction heating in the motor industry. Extremely quick heating times are possible, a very high degree of consistency can be obtained, and there is only a low percentage of scrap.—U. E.

ESA—A New Oxygen-Arc Cutting Process. E. Brünn and F. Erdmann-Jesnitzer. (*Schweisstechn.*, 1953, **7**, June, 68–70). The ESA oxygen-arc cutting method differs from the already known oxy-arc method in that the oxygen stream hits the surface directly behind the arc. A description is given of electrodes and electrode holders.—U. E.

Recent Developments in Powder Processes. R. S. Babcock. (*Canad. Metals*, 1953, **16**, July, 42–47). Although the powder cutting process was originally developed for cutting stainless steel it now has a much wider field of application. The cutting lance consists of a mechanism which introduces iron powder as a stream into an oxygen jet. The use of this lance for cutting heavy scrap, skull reduction, and opening iron and cinder notches in blast-furnaces is described.—J. C. B.

MACHINING AND MACHINABILITY

New Concepts Regarding the Machinability of Materials. G. Bendiksen. (*Dansk Tekn. Tidsskr.*, 1953, **77**, Feb., 31–35 ; Mar., 59–65 ; Apr., 98–100). Advantages and failings of test methods for checking tool quality or machinability of metals in relation to optimum cutting speed are discussed. The cutting pressure is shown to be of paramount importance. The best method of improving machining characteristics is to reduce friction. Formulae are presented which enable minimum machining costs to be determined.—R. A. R.

Correlation of Machinability and Other Properties of Materials. F. Rapatz. (*Stahl u. Eisen*, 1953, **73**, Aug. 27, 1169–1174). Insufficient distinction is made in practice between tool life and appearance of the machined surface when discussing machinability. Recent work shows that, even with similar materials, tool life cannot be determined from tensile strength. The relationship between log (tool life) and log (cutting speed) cannot, in general, be extrapolated to longer or shorter times than those for which the relationship was established. Surface appearance of machined articles is influenced by the type of chip. Increase of cutting speed improves the surface quality ; an improvement can also be made by heat treating the material to produce brittleness or an increase in tensile strength. In this connection, the good machinability of free cutting steels, with respect to tool wear and surface appearance, is to be attributed to the lubricating action of the inclusions, which reduce the coefficient of friction, and to the greater brittleness of the steel. Of the accelerated tests, that in which tool life is measured in terms of length of cut offers the best prospect of success for practical assessment of machinability.—J. P.

Surface Roughness and Its Testing. F. Motalik. (*Österr. Masch. Elektrowirtschaft*, 1953, **8**, July, 187–190). The author first summarizes the methods of surface treatment and then surveys procedures for testing surface roughness, including needle scanning and optical methods.—L. D. H.

How Are the Surfaces of Materials Tested ? F. Gabler. (*Österr. Masch. Elektrowirtschaft*, 1953, **8**, Mar., 68–70). Testing the uniformity of metallic and other surfaces by measuring interference patterns is described, the principles are outlined, and the use of the Reichert apparatus for this purpose is discussed.—L. D. H.

Grinding Carbide Tools with Electrolytic Assistance. N. W. Thibault and B. H. Anderson. (*Metal Progress*, 1953, **64**, Aug., 161–166). A description is given of a new technique of grinding carbide type materials. The conventional diamond wheel is employed together with the controlled erosion of the carbide by direct current methods.—B. G. B.

CLEANING AND PICKLING

Notes on Pickling and Descaling. A. Wogrinz. (*Österr. Masch. Elecktrowirtschaft*, 1953, **8**, July, 190–192). A short account is given of the cause of scale formation, its composition, and chemical and electrical methods of scale removal.—L. D. H.

Electrolytic Pickling of Iron and Steel in Alkaline Solutions and Melts. R. Springer. (*Metalloberfläche*, 1953, **7**, Jan., A9–A12). A survey is made of methods in use or available for pickling with anodic, cathodic or alternating polarity, using alkaline solutions or fused salts. Some working data are given, together with an outline of the principles involved. (22 references).—L. D. H.

Electrolytic Polishing. W. Pfanhauser. (*Österr. Masch-Elektrowirtschaft*, 1953, **8**, July, 192–193). Methods of elec.

trolytic polishing are summarized, and the importance of correct electrical and temperature conditions, and electrolyte composition are emphasized.—L. D. H.

Stripping Phosphate Coatings.—Radiometric Evaluation of a New Stripping Solution. S. L. Eisler and J. Doss. (*Metal Finishing*, 1953, **51**, Aug., 58–63). Iron and zinc phosphate deposits on steel can be removed by immersion for 10 min. at room temperature in a solution containing sodium hydroxide 24 oz., sodium cyanide 12 oz. and tetrasodium ethylenediaminetetracetate 12 oz. per gallon. Manganous phosphate coatings require the work to be made cathodic (c.d. 50 amp./sq. ft.) for 10 min. in a bath at 160° F. Stripped articles can be readily re-phosphated.—J. P.

The Stripping of Plating Shop Rejects. C. D. Leonard. (*Electroplating*, 1953, **6**, Aug., 290–292, 296). Some general considerations are given for the stripping of rejects. These include choice of solution, use of electrolytic or chemical polishing methods, lay-out of equipment, storage and cleaning of rejects before stripping, and control. Some mechanical processes are also described.—J. P.

The Problem of Hydrogen Diffusion in the Pickling of Spring Steel. J. S. Jackson. (*J. Electrodep. Tech. Soc.*, 1952, **28**, 89–101). The author discusses the effect of occluded hydrogen on the embrittlement of spring steel. Acid or cathodic pickling is most likely to produce occluded hydrogen, the influence of which is most pronounced in hardened and tempered materials. Although the hydrogen can be expelled to a large extent by heat-treatment, this procedure is never fully satisfactory and elimination of pickling or the use of special pickling treatments is to be preferred.—J. P.

Coils Spin Through Rotary Pickler. (*Steel*, 1953, **133**, Sept. 28, 92–94). A low cost pickling installation for coiled strip is available, using a new operating principle. Coils up to 24 in. wide can be cleaned and oiled at over 3 tons/hr. Coils are treated in cages which are rotated in the tanks, causing the coils to unwind and rewind.—D. L. C. P.

PROTECTIVE COATINGS

Factors Influencing the Design of Automatic Plating Plant. A. Smart. (*J. Electrodep. Tech. Soc.*, 1951, **27**, 277–291). The development of an automatic plant for degreasing, pickling, zinc plating and phosphating welded steel window frames from the laboratory stage through the pilot plant to the commercial scale is described.—J. P.

Tanks for Electroplating and Pickling. F. Spicer. (*Mech. World*, 1953, **133**, June, 266–268). A brief survey is given of the materials used in the construction and lining of electroplating and pickling baths. Mention is made of the various methods employed for heating such baths.—D. H.

Theoretical Principles of Electroplating Practice. H. Wagner. (*Metallwaren-Industrie und Galvanotechnik*, 1953, **44**, Nov.–Dec., 64–88). This paper is intended as an introduction to the elements of physical chemistry and electrical theory for electroplating operators with no theoretical knowledge of these subjects. The fundamentals of electrolytic action are outlined, and the laws of electrolysis are briefly discussed.

Protective Coatings on Metals. D. M. Dovey and K. C. Randle. (*Metal Treatment and Drop Forg.*, 1953, **20**, Aug., 341–345). The important part played by metallic and ceramic surface coatings in increasing the life of metals and alloys at high temperatures, particularly in the presence of fuel-ash constituents in high-velocity gas streams, is discussed. Vapour deposition of metals, carbides, borides, and silicides is reviewed, and the ductility and oxidation resistance of such coatings are tabulated. Developments in the use of ceramic coatings and in cladding, sintering, and hot-dipping are outlined.—P. M. C.

Use and Economics of Metallic Coatings. H. W. Dettner. (*Metall*, 1953, **7**, July, 552–554). Methods of applying metallic coats are enumerated, their uses indicated and costs and profit are referred to.—J. G. W.

Control of Electric Current and Related Variables in Plating and Anodizing. (*Electroplating*, 1953, **6**, July, 235–244 ; Aug., 285–289). The importance of current control in electroplating is stressed and the various methods, manual and automatic, by which this may be done are described and illustrated. The possibility of linking current control with that of other variables, such as ampere-hours, temperature and programme control is discussed.—J. P.

***p*H—Its Meaning, Significance, and Control in Plating Baths.** F. Spicer. (*Electroplating*, 1953, **6**, May, 160–166). An elementary exposition is given of the meaning of the term *p*H and methods and instruments for measuring it are described and illustrated.—J. P.

Density and Direction of the Current Paths in Plating Baths. J. Steiner and M. Ruiner. (*Metalloberfläche*, 1953, **7**, May, B69–B72). By the use of a double copper probe, the construction and operation of which are described, the current density and direction of the current field is mapped out for several types and sizes of electrodes, with and without the use of diaphragms. The use of this probe is compared with that of a ring cathode for the same purpose.—L. D. H.

Anodes for Electroplating. A. C. West. (*Metal Finishing*, 1953, **51**, Aug., 66–70, 74). The composition and form of copper, cadmium, zinc, and nickel anodes for electroplating are discussed.—J. P.

Current and Metal Distribution in Electrodeposition. III—Experimental Determination of Metal Distribution. J. Kronsbein. (*Plating*, 1953, **7**, Aug., 898–901). Results obtained by measuring the metal distribution over the interior surfaces of right angle specimens of varied shank length and fillet radius plated in a specially designed shielded box in purified copper sulphate, Rochelle type copper cyanide, and Watts' nickel baths show that fillet radius has very little effect on the thickness of deposit in corners and along the adjacent shanks when such radius exceeds about 0·02 in. per 2 in. of shank length. Designers should try not to go below this figure.—J. P.

Plating and Painting Steel Shell Cases. (*Product Finishing*, 1953, **6**, July, 96–98). Current practice in the U.S.A. in the manufacture of steel cartridge cases and their plating and spray-varnishing are briefly described.—J. P.

Economy in Metal Finishing. S. Wernick. (*Trans. Inst. Met. Finishing*, 1953, **29**, Advance Copy No. 12). The methods by which economies may be effected in the preparation of work and the electrodeposition of metal finishes are discussed. The review stresses removal of conditions which produce defects.—J. P.

Bright Metal Surfaces and a Theory of Bright Plating. T. P. Hoar. (*Trans. Inst. Met. Finishing*, 1953, **29**, Advance Copy No. 11). The author discusses the nature of a bright metal surface and shows that the production of such a surface, except by mechanical means or rupture at cleavage planes, requires either random deposition or removal of metal atoms. Such a requirement is met in electroplating by the use of addition agents and the author suggests compounds which may be useful as brighteners.—J. P.

The Nature of the Film Present on Iron after Brightening in Marshall's Solution. A. Hickling, W. A. Marshall, and E. R. Buckle. (*J. Electrodep. Tech. Soc.*, 1952, **28**, 47–68). The presence of a film upon iron after treatment in the Marshall smoothing solution (oxalic acid dihydrate 2·5 g., H_2O_2 1·3 g., H_2SO_4 0·01 g. per 100 ml.) has been demonstrated. Evidence indicates that this film consists of a very thin but tenacious and continuous layer of iron oxide, the thickness of which has been estimated by cathodic reduction to be of the order of 60Å. A similar film rapidly reforms when a cathodically stripped specimen is exposed to air.—J. P.

Carbide Flame-Plating Presents New Possibilities. W. L. Donnelly. (*Steel Processing*, 1953, **39**, June, 279–282). The flame deposition of tungsten carbide hard-metal on to steel and many other metals is described. Coatings from 0·0005 to 0·020 in. thick can be applied where wear resistance is required. Many examples of the application of the process are described. In applications where thermal shock is likely to be encountered, thin coatings should be used in preference to thick ones.—P. M. C.

Plating Through a Slot. J. B. Mohler. (*Metal Ind.*, 1953, **83**, Sept. 11, 219–222). The distribution of current over a cathode, and the effect of anode position and proximity of plating bath walls are discussed. Methods of producing uniformity and nonuniformity by enclosing the cathode in a box with open end or with narrow slot are detailed. The use of such slot-cells as plating-range cells for addition agent control of plating baths is discussed.—P. M. C.

Factors Influencing the Selection of Metal Finishes. A. W. Wallbank. (*Sheet Metal Ind.*, 1953, **30**, July, 585–588, 590). The purposes of metal finishing, and the advantages of electrodeposition are discussed. Anodic and cathodic deposition are described, and the service performance, uses, appearance, and cost of the commonly available deposits are outlined.

A Survey of Modern Industrial Finishes for Sheet Metal Products. H. J. Testro. (*Sheet Metal Ind.*, 1953, **30**, Aug., 665–668, 680). The author outlines certain of the principal types of organic finishing materials, and discusses their specific properties and applications. Plain finishes considered are primers, celluloses, 'half-hour' enamels, alkyd resins, vinyl resins, styrenated alkyds, chlorinated rubber paints and silicone finishes. Polychromatic, hammer, and wrinkle finishes for decorative use are dealt with.—P. M. C.

Modern Magnetic Thickness Testers. (*Electroplating*, 1953, **6**, July, 249–255). The principles and modes of operation of some commercial permanent and electromagnetic layer thickness meters are described. The limitations and fields of application of each instrument are indicated.—J. P.

Position of the Standardization of Commercial Surfaces. W. Schenkel and W. Schmidt. (*Metalloberfläche*, 1953, **7**, Jan., A1–A6). The modern DIN (German Standard) surface specifications are explained, together with a code system and illustrative data. The German definitions are compared with the British and American conventions.—L. D. H.

An Interferometer for Examining Polished Surfaces. R. E. Sugg. (*Mech. Eng.*, 1953, **75**, Aug., 629–631). The author discusses the principle and application of multiple beam interferometry. A portable interferometer is described designed for use as a works instrument.—D. H.

Surface Measurements with Contact Scanning Instruments. W. Schmidt. (*Metalloberfläche*, 1953, **7**, Mar., A33–A38). The principles of this class of instrument are first outlined, and a number of typical instruments from English, American, and German practice are described and illustrated, and lines of future development are suggested.—L. D. H.

Plating Waste Treatment. C. F. Paulson. (*Metal Progress*, 1953, **64**, Aug., 84–87). The treatment of electroplating wastes by ion exchange methods is discussed and examples are given of suitable plant and of the costs.—B. G. B.

The Significance and Treatment of Cyanide in Industrial Waste Waters. E. A. Whitlock. (*Electroplating*, 1953, **6**, Sept., 325–335). Of the numerous methods investigated for removing cyanide from waste waters, only two have achieved any industrial importance, viz., the ferrous-sulphate/lime and the alkaline chlorination processes. Moreover, of these two, only chlorination is capable of producing an effluent suitable for discharge to a water course without endangering fish life or to public sewers where zero cyanide content is specified.

Disposal of Plating Room Wastes. VI—Treatment of Plating Room Waste Solutions with Ozone. C. A. Walker and W. Zabban. (*Plating*, 1953, **7**, July, 777–780). The oxidation of free and complex cyanides by ozone results in the formation of free cyanates, insoluble metal cyanates and hydroxides. Reaction is very rapid. Oxidation does not proceed beyond the cyanate stage, and must be catalysed by oxygen carriers such as copper or manganese. In absence of catalysts, a large proportion of the ozone leaves the solution unreacted. Within the range 7·0–12·5, *p*H appears to have little effect on the oxidation. Ferrocyanides are not decomposed by ozone.—J. P.

Occupational Eczema in Electroplating Operations, and Its Prevention. W. Kohl. (*Metalloberfläche*, 1953, **7**, Jan., B1–B2). The symptoms and causes of eczema contracted in galvanizing plants are discussed; a comprehensive list of the substances likely to cause trouble is given, and the author summarizes preventive measures, including a number of protective creams which are available in Germany.—L. D. H.

Diseases Caused by Chromium, Cadmium, and Nickel to Workers in Plating Plants. B. Pernis. (*Ingegneria Meccanica*, 1953, **2**, July, 31–32). [In Italian]. A brief discussion is given on the diseases to which workers in plating plants are subject. Their causes are described, and precautious which should be taken are discussed.—M. D. J. B.

Chromium Plating for 0·50 in. Gun Barrels. (*Canad. Metals*, 1953, **16**, July, 38–39). The barrels of the 0·50 in. machine guns are partly chromium plated to increase wear resistance and to produce a tapered bore. This article describes the plating installation of Canadian Arsenals Ltd., Long Branch. Lead electrodes are used for the electrolytic polishing before the plating operation as well as during actual plating; they consist of 0·60%-carbon steel rods plated with a 7% tin-lead alloy. This coating is work hardened by drawing the rod through a die.—J. C. B.

Stability of Ion Exchange Resins toward Chromic Acid Plating Solutions. C. F. Paulson and G. H. Saunders. (*Plating*, 1953, **7**, Sept., 1005–1009). Methods of employing ion exchange resins to remove metal contaminants from chromium plating baths so as not to degrade the resins are outlined.—J. P.

Crack-Free Chromium from Baths of the Self-Regulating High-Speed Type. R. Dow and J. E. Stareck. (*Plating*, 1953, **7**, Sept., 987–996). Crack-free chromium plate, deposited from a proprietary solution, shows promise of having a wide variety of applications. Because of its freedom from porosity, it has better corrosion resistance than conventional chromium plate.—J. P.

Black Chromium-Base Electroplating. M. F. Quaely. (*Plating*, 1953, **7**, Sept., 982–986). A hard bright chromium base electrodeposit has been developed which can be applied at a high rate of deposition. The electrolyte contains chromic acid, nickel chloride and acetic acid. When this deposit is treated with hydrochloric acid for a few seconds, a black finish results, which is non-smudging, very adherent and uniform. Another black deposit has been developed which is deposited from an electrolyte containing chromic acid, nickel chloride, and vanadium nitrate and acetic acid. The coatings have good heat and chemical resistance.—J. P.

Chromizing Improves Surface Properties of Steels. R. P. Seelig. (*Mat. Methods*, 1953, **37**, May, 106–109). The 'Chromalloy' process is described, in which a surface layer rich in chromium or chromium carbides is produced by heating steel in a patented powder compound. The chromium content of the case is about 24%, the thickness being dependant on surface condition, type of steel, and time and temperature of treatment. Properties of the case are discussed.—P. M. C.

Doubled Life between Regrinds by Means of Hard Chromium Plating. R. Brunneder. (*Osterr. Masch. Elektrowirtschaft*, 1953, **8**, July, 201–202). The advantages of hard and bright chromium plating in giving increased hardness and resistance to corrosion and wear are described. The increased life obtained by hard chromium plating tools and saw blades is shown by examples.—L. D. H.

Hard Chromium Plating. K. Schwartzböck. (*Osterr. Masch. Elektrowirtschaft*, 1953, **8**, July, 198–201). A general survey is given of the hard chromium plating process. Bath compositions, current strength, suitable current densities, mechanical details, pre- and after-treatment of the metal, and porous chromium plating are discussed.—L. D. H.

Black Nickel Plating. (*Australasian Plating Fin.*, 1952, **2**, Oct.–Nov., 10–11). Some solutions and operating procedures for obtaining black nickel deposits are briefly described.—J. P.

Bright Nickel Plating. W. Pfanhauser. (*Osterr. Masch. Elektrowirtschaft*, 1953, **8**, July, 196–198). After indicating the industrial importance of nickel plating, and dealing briefly with the incidence and testing of porosity in nickel coatings, the development of bright nickel plating is described, followed by a discussion of the principles.—L. D. H.

Electroless Nickel Plating: Where it Stands Today. J. B. Campbell. (*Mat. Methods*, 1953, **37**, May, 96–100). Details of an entirely new nickel plating process, which makes use of chemical reaction instead of electrolytic action, are given. The so-called nickel plate is a nickel phosphorus alloy containing 90 to 95% nickel. The original process is known as 'hustralloy', but a modification known as the 'Kanigen' process has recently been developed by General American Transportation Corp. The chief advantages of this 'electroless' plate compared with electrodeposits are its applicability to many parts which cannot be successfully electroplated, better corrosion protection for a given thickness, and less nickel required to insure a given minimum plate thickness.

Electroless Plating Produces Hard Nickel Plating. E. L. Gostin. (*Iron Age*, 1953, **171**, June 11, 115–119). **Electroless Plating.** E. L. Gostin. (*Metal Ind.*, 1953, **83**, July 17, 47–48). The chemical reduction of nickel from solutions of its salts can be controlled to give hard, adherent deposits on a variety of base metals. Process requires no electric power and intricate shapes can be plated uniformly. Low temperature heat-treatment improves the hardness, and the wear and corrosion resistance. No details of the solution are given but at a bath temperature of 200° F. the plating rate is about 0·0005 in./hr.—A. M. F.

The Diffusion of Iron in Nickel. M. B. Neiman, A. Ya. Shinyaev, and B. G. Dzantiev. (*Doklady Akademii Nauk S.S.S.R.*, 1953, **91**, (2), 265–267). [In Russian]. The diffusion

of iron in nickel in the temperature range 950–1130° C. was studied using radioactive Fe_{59}. Thin nickel foil was plated on one side with iron containing radioactive Fe_{59}, placed in a silica vessel and heated *in vacuo* to a given temperature. The rate of diffusion was followed by measurements of the radioactivity on both sides of the foil. The temperature dependence of the diffusion coefficient of iron in nickel and the activation energy were determined. The velocity of the diffusion of iron into nickel is considerably higher than that of nickel into iron.—V. G.

Formation of " Tears " on Galvanized Sheets. H. J. Wiester and D. Horstmann. (*Stahl u. Eisen*, 1953, **73**, July 2, 906–908). Metallographic examination has shown that retraction or de-wetting after hot galvanizing is due to oxide residues on the surface of steel sheets.—J. P.

POWDER METALLURGY

The Stability of Inorganic Compounds in High Vacuum. M. Auwärter. (" *Pulvermetallurgie*," *First Plansee Seminar* " *De Re Metallica* ", June 22–26, 1952, Reutte, 1–7). [In German]. Processes occurring during high-vacuum melting of inorganic compounds, particularly oxides of zinc, nickel, iron, and copper, are described. The dissociation pressure of oxygen and reactions with the environment play decisive parts. Conclusions are drawn concerning high-vacuum melting and sintering of metals and their alloys and compounds.—K. E. J.

The Influence of Adsorbed Gases on the Stability of Crystal Lattices near the Surface. H. Forestier, J. Maurer, J. P. Kiehl, and P. Stahl. (" *Pulvermetallurgie*," *First Plansee Seminar* " *De Re Metallica*," June 22–26, 1952, Reutte, 8–20). [In French]. Gas adsorption produces changes in the reactivity of solid oxides, transformation temperature of powdered polymorphous substances, m.p. of crystalline substances, and speed of transformations (*e.g.*, in martensite and beryllium bronze). Experimental data are interpreted theoretically : the gas, adsorbed by Van der Waal's forces, alters the stability of energy levels in the solid surface. (13 references).—K. E. J.

Micro-Thermal Analysis, Melting and Crystallization of Solid Solutions. A. Kofler. (" *Pulvermetallurgie*," *First Plansee Seminar* " *De Re Metallica*," June 22–26, 1952, Reutte, 21–26). [In German]. Microfilm techniques can be used to demonstrate : Formation of eutectics, molecular compounds, miscibility gaps in liquid phases, and solid solution in binary systems ; and homogenization of solid solutions on heating and cooling in the solidus/liquidus interval. Homogenization depends on periodical transformations, and the presence of small amounts of melt when produced in a few hours. (17 references).—K. E. J.

A Uniform Hardness Testing Method (Penetration Hardness) Using Hard Metals. R. Mitsche. (" *Pulvermetallurgie*," *First Plansee Seminar* " *De Re Metallica*," June 22–26, 1952, Reutte, 27–38). [In German]. Extensive experiments show that a uniform hardness scale (ranging from pure aluminium to fully-hardened steels) can be based on indentation tests with a 136° pyramid of sintered hard metal. The evaluation is based on depth of penetration or on calculation of the Vickers hardness number. One load only (150 kg.) need be used. (15 references).—K. E. J.

The Chemical Bond in Monocarbides and Its Relation to Hardness. H. Nowotny and F. Vitovec. (" *Pulvermetallurgie*," *First Plansee Seminar* " *De Re Metallica* ", June 22–26, 1952, Reutte, 39–48). [In German]. Present theories of bonding in lattices of intermetallic phases are reviewed. Series of isomorphous compounds are discussed, and different types of bonding are distinguished. Hard metal carbides are considered in relation to other compounds containing two atoms only in the simplest unit. The existence of simple relationships between type of bonding and properties such as hardness is discussed. (25 references).—K. E. J.

Some New Concepts of the Strength of Alloys. G. M. Schwab. (" *Pulvermetallurgie*," *First Plansee Seminar* " *De Re Metallica*," June 22–26, 1952, Reutte, 49–64). [In German]. The temperature coefficient of the Brinell hardness of metals is represented by a bi-exponential formula, which implies two processes with different activation energies : The larger, E_2, is connected with formation of new centres of activity. The hardness of the α, β, γ, ϵ, and η phases of Cu–Zn, Cu–Sn, Ag–Zn and Ag–Cd were measured and evaluated : E_2 shows a maximum in all γ-phases, and this explains the hardness maxima at room temperature.—(22 references).

K.E.J.

Diffusion in Solid Metals. W. Seith. (" *Pulvermetallurgie*," *First Plansee Seminar* " *De Re Metallica*," June 22–26, 1952, Reutte, 65–80). [In German]. New experimental results are reviewed from the field of diffusion in metals (particularly the systems Fe–Ni, Fe–C–Ni, silicon steel, manganese steel, Ag–Au, and Cu–Zn). The modern concept of the interchange mechanism, as derived from these results, is discussed. Particular attention is given to conditions during sintering of metal powders. (16 references).—K. E. J.

Present Theories of the Metallic State and Their Importance in Metal Ceramics. (Powder Metallurgy). F. Skaupy. (" *Pulvermetallurgie*," *First Plansee Seminar* " *De Re Metallica*," June 22–26, 1952, Reutte, 81–88). [In German]. Present views on the nature of metals are reviewed, with particular reference to interatomic distances and the properties of thin layers. These concepts are applied to some problems of powder metallurgy.—K. E. J.

Ideal and Real Crystals. A. Smekal. (" *Pulvermetallurgie*," *First Plansee Seminar* " *De Re Metallica*," June 22–26, 1952, Reutte, 89–99). [In German]. Microstructures of real crystals contain ideal structure elements separated by deficiencies, and are influenced by conditions of crystallization, lattice forces, and impurities. The sizes of the ideal portions vary greatly with different substances and the dimensions may be determined by light scratching. In several cases (including WC) the dimensions are considerably above the light-optical dissolution point. (16 references).—K. E. J.

Fundamentals of Wear in Solids Under Dry Conditions. R. Walzel. (" *Pulvermetallurgie*," *First Plansee Seminar* " *De Re Metallica*," June 22–26, 1952, Reutte, 100–105). [In German]. Wear is due to friction forces exceeding the tensile and shear strength of the surface layer. It is difficult to predict wear resistance, and to extrapolate laboratory results to service conditions, since friction force depends on many factors and the surface tensile and shear strengths are affected by changes taking place during wear. An attempted correlation of the factors involved is given, with special reference to sintered materials.—K. E. J.

Some Observations on the Mechanism of Liquid Phase Sintering. H. S. Cannon and F. V. Lenel. (" *Pulvermetallurgie*," *First Plansee Seminar* " *De Re Metallica*, " June 22–26, 1952, Reutte, 106–122). [In English]. As a result of studies of sintering conditions, pressure, and powder properties in the systems Fe–Cu, Cu–Ni, Cu–Bi, Cu–Pb, Ag–Pb, and W–Cu, a theory of liquid phase sintering is advanced. Rapid partial densification occurs initially, due to a flow mechanism of liquid into the pores of the compact. Cessation of densification short of theoretical density is due to formation of a rigid skeleton of solid phase, but growth of solid particles by solution and reprecipitation may then continue.—K. E. J.

The Development of High Strength, Heat-Treatable Products from Alloy Powders. G. J. Comstock and F. H. Clark. (" *Pulvermetallurgie*," *First Plansee Seminar* " *De Re Metallica*," June 22–26, 1952, Reutte, 123–139). [In English]. After a discussion of economic and other advantages of powder metallurgical applications, the production of parts from alloy steels and cobalt-base alloys is discussed. Particular attention is given to porous parts, particle shape and permeability.—K. E. J.

Determination of the Adhesive Force of Metal Powders. E. Cremer. (" *Pulvermetallurgie*," *First Plansee Seminar* " *De Re Metallica*," June 22–26, 1952, Reutte, 140–145). [In German]. A characteristic value termed ' adhesive force ' may be determined by measuring the glide angle of a powder dusted on a solid support. It depends on the surface properties of both powder and support, and is inversely proportional to the particle diameter of the powder.—K. E. J.

The Effect of Lattice Changes on the Sintering Process. H. H. Hausner. (" *Pulvermetallurgie*," *First Plansee Seminar* " *De Re Metallica*," June 22–26, 1952, Reutte, 146–156). [In English]. Sintering studies using nickel, zirconium, and ZrH_2 show that lattice changes affect the sintering process. In some cases it may be more advantageous to use metallic compound powders which decompose during sintering. The presence of nascent hydrogen is assumed to be favourable to the sintering process.—K. E. J.

On the Separation of Chemically Pure Vanadium from the Gaseous Phase. G. Jantsch and L. Zemek. ("*Pulvermetallurgie*," *First Plansee Seminar* "*De Re Metallica*" June, 22–26, 1952, Reutte, 157–159). [In German]. The method is based on hydrogen reduction of VCl_4, and yields a product containing 99% V, balance hydrogen. Properties of the particles and lattice are given. (11 references).—K. E. J.

The Production of Metal Powders by Fusion Electrolysis. W. J. Kroll. ("*Pulvermetallurgie*," *First Plansee Seminar* "*De Re Metallica*," June 22–26, 1952, Reutte, 160–170). [In English]. Metal powders can be produced by fusion electrolysis at temperatures where the cathode deposit appears as small crystals. The method may be used for powder winning, metal purification or producing alloy powders. The crystals show advantages of purity and structure. Current work on flake titanium should give valuable information on cell operation. Techniques for producing tungsten, molybdenum, manganese, and alloy powders are specially discussed.—K. E. J.

Powder Metallurgy as Viewed by a Manufacturer of Metal Powder Products. A. J. Langhammer. ("*Pulvermetallurgie*," *First Plansee Seminar* "*De Re Metallica*," June 22–26, 1952, Reutte, 171–174). [In English]. Production of ferrous and non-ferrous heavy-duty self-lubricating bearings and finished machine parts is discussed. Many advantages of powder metallurgical methods are discussed. and case histories given. The need for raw materials of uniform quality is stressed.—K. E. J.

Microhardness as an Aid to Understanding Sintering Processes in Complex Systems. E. M. Onitsch-Modl. ("*Pulvermetallurgie*," *First Planee Seminar* "*De Re Metallica*," June 22–26, 1952, Reutte, 175–187). [In German]. The microhardness test can be used as a simple means of examining alloying processes which incorporate hardness changes. It supplements normal metallographic tests. Diffusion processes in the matrix and during carbide formation were followed during the sintering of iron–carbon alloys containing chromium, tungsten, molybdenum, and vanadium. Applications and limits of the test are discussed. (11 references).—K. E. J.

A Contribution to the Physical Analysis of Sintering Processes. G. Ritzau. ("*Pulvermetallurgie*," *First Plansee Seminar* "*De Re Metallica*," June 22–26, 1952, Reutte, 188–198). [In German]. The course of sintering may be followed by direct measurement of electrical conductivity and thermoelectric potential. The former varies with porosity snd diffusion changes in microstructure ; the latter is independent of porosity. The method is discussed for systems exhibiting mutual solubility (copper–nickel) and insolubility (copper–iron). The method will follow sintering in complex alloys (hard metals and magnet materials).—K. E. J.

Surface Layers, Particularly Oxide Layers, and Their Influence on Sintering. F. Sauerwald. ("*Pulvermetallurgie*," *First Plansee Seminar* "*De Re Metallica*," June 22–26, 1952, Reutte, 199–202). [In German]. The stability of surface layers is important when assessing their effects on sintered metals. The effects of stable and unstable oxide films on the mechanical strength of copper compacts are discussed. Under certain conditions, surface oxide films may have a favourable effect on compact strength.—K. E. J.

Applications of Radioactive Techniques in Powder Metallurgy. H. Schreiner. ("*Pulvermetallurgie*," *First Plansee Seminar* "*De Re Metallica*", June 22–26, 1952, Reutte, 203–210). [In German]. Surface or lattice activation, using suitable isotopes, provide a means for studying sintering behaviour and surface and lattice diffusion, giving clear results at all temperatures, including those of sintering itself. Examples are given of applications to iron, copper, and the iron–copper system.—K. E. J.

A System of Classification of Hard Metals. C. Ballhausen. ("*Pulvermetallurgie*" *First Plansee Seminar* "*De Re Metallica*", June 22–26, 1952, Reutte, 221–231). [In German]. Properties of WC–TiC–Co hard metals are plotted as space co-ordinates on a base whose abscissa is the volume ratio WC/TiC and ordinate the cobalt content. Compressive strength, modulus of elasticity, hardness, modulus of rupture, coefficient of expansion, and cutting characteristics are so treated. Specifications may then be framed in terms of WC/TiC ratio and cobalt content.—K. E. J.

The Solid Solubilities of Some Stable Carbides in Cobalt, Nickel and Iron at 1250° C. R. Edwards and T. Raine. ("*Pulvermetallurgie*," *First Plausee Seminar* "*De Re Metallica*", June 22–26, 1952, Reutte, 232–243). [In English]. In view of the use of other materials than tungsten and cobalt for hard metal production, for strategic purposes and to achieve improved properties, other carbide-forming and base metals have been investigated. Solubilities and crystal and phase structure details are given of WC, TiC, WC/TiC, TaC, NbC, Mo_2C, V_4C_3, and Cr_3C_2 in cobalt, nickel and iron.

High-Temperature-Resisting Materials by Siliconizing Tungsten and Molybdenum. E. Fitzer. ("*Pulvermetallurgie*," *First Plansee Seminar* "*De Re Matellica*", June 22–26, 1952, Reutte, 245–258). [In German].

Preparation of Pure Titanium Carbide. G. F. Hüttig. ("*Pulvermetallurgie*," *First Plansee Seminar* "*De Re Metallica*", June 22–26, 1952, Reutte, 259–267). [In German].

Solid Solution Formation in High-Melting-Point Hard Metal Compounds. R. Kieffer. ("*Pulvermetallurgie*," *First Plansee Seminar* "*De Re Metallica*", June 22–26, 1952, Reutte, 268–296). [In German]. The term 'hard metals' refers to carbides, nitrides, borides, and silicides of the transition metals of the fourth to sixth groups in the Periodic Table. Solid solubilities of the isomorphous carbides in the fourth and fifth groups, and partial solubilities between these and those in the sixth group, are discussed. Nitrides in the fourth and fifth groups mostly form solid solutions with each other and the isomorphous carbides. Isomorphous bodies and silicides also tend to form solid solutions ; these are being used for high-temperature materials. Carbides and nitrides generally have simple structures : borides and silicides form many phases of complicated structure. Diffusion processes in solid solution may be followed by metallographic and microhardness methods. (42 references).—K. E. J.

Investigations on Ternary Systems Me_1–Me_2–B; and a Discussion of the Relative Strength of the Bond Transition Metal-Boron. R. Kiessling. ("*Pulvermetallurgie*," *First Plansee Seminar* "*De Re Metallica*", June 22–26, 1952, Reutte, 297–302). [In English]. Details are given of distribution ratios, etc., in boride systems containing manganese, iron, cobalt, and nickel.—K. E. J.

The System Tungsten–Cobalt–Carbon. P. Rautala and T. Norton. ("*Pulvermetallurgie*," *First Plansee Seminar* "*De Re Metallica*", June 22–26, 1952, Reutte, 303–316). [In German]. Phases and equilibria in the system have been studied by X-ray diffraction and metallographic and thermal techniques. As well as η-phases, two double carbide phases ($Co_3W_6C_2$ and $Co_3W_{10}C_4$) have been revealed, with face-centred cubic and hexagonal lattices. Isomorphous phases form in corresponding iron and nickel systems. Reactions during phase formation and sintering, and equilibrium diagrams, are discussed.—K. E. J.

PROPERTIES AND TESTS

Testing the Hardenability of Constructional Steel. H. Kiessler. (*Härterei-Technische Mitteilungen*, 1950, **5**, 97–108). Hardenability is first defined, and the factors which influence this property are assessed. The principal testing methods are described, and results obtained on specific samples are discussed.—L. D. H.

Finotest—The New Hardness Tester for Small Loads. (*Microtechnic*, 1953, **7**, (3), 155–156). This indentation instrument, manufactured by Karl Frank Ltd., is suitable for loads of 30 to 10,000 g., and is intended for the measurement of thin test pieces down to 0·03 mm. It can be used for scratch as well as indentation tests, and its principal advantages are accurate location of the test position measurement to 0·005 mm., simple and rapid adjustment, and a photographic attachment.—L. D. H.

Strain Ageing in Iron and Steel. J. D. Fast. (*Osterr. Masch. Elecktrowirtschaft*, 1953, **8**, Jan., 12–16 ; Feb., 40–43). After describing the phenomenon of strain ageing, and its effects on the properties of steel, the author explains the influence of carbon and nitrogen, particularly the latter, in promoting ageing. The anomaly in the behaviour of manganese is discussed.—L. D. H.

Influence of Aluminium and Silicon Deoxidation on the Strain Ageing of Low-Carbon Steels. W. C. Leslie and R. L. Rickett. (*Trans. Amer. Inst. Min. Met. Eng.*, 1953, **197**: *J. Met.*, 1953, **5**, Aug., 1021–1031). The authors have studied the effects of composition, heat-treatment, and ageing conditions on strain ageing of samples from two low-carbon steel heats, one silicon-killed and one rimming, with graded

aluminium additions made to each. Strain ageing, primarily due to nitrogen in solution in ferrite, is reduced by silicon and aluminium due to the formation of nitrides, the two elements together being more effective than either element alone. Heat-treatments which tend to precipitate nitrides also reduce subsequent strain ageing.—G. F.

Rapid Calculation of Microhardness Numbers. L. E. Samuels and T. O. Mulhearn. (*Metal Progress*, 1953, **63,** May, 82–83). The calculation of microhardness numbers is often laborious. A description is given of two slide-rule methods which combine the multiplication constants used in the calculation. Details are also given of the construction and use of a special microhardness calculator.—B. G. B.

Effect of Grain Size upon Temper Brittleness. L. D. Jaffe, F. L. Carr, and D. C. Buffum. (*Trans. Amer. Inst. Min. Met. Eng.*, 1953, **197 :** *J. Met.*, 1953, **5,** Sept., 1147–1148). The authors have studied the effect of austenitic grain size on temper brittleness in a low-alloy Ni–Cr steel. They conclude that increasing grain size increases the shift in the transition temperature associated with temper brittleness, and that, at fixed grain size, a decrease in quenching temperature has little effect on the transition temperature.—G. F.

Losses in Iron Laminations. (*Elect. Rev.*, 1953, **153,** Aug. 14, 335–336). A method is described for measuring energy losses in laminated iron cores at high flux densities. By the use of thermocouples connected to the specimen and a guard strip of Nichrome, and by heating the guard strip separately to the same temperature as the specimen to avoid heat transference losses, the iron loss is determined from the energy supplied to the Nichrome, the specific heats of the two materials being known. B_{max} values of up to 23,750 gauss were obtained.—L. D. H.

The Surface Tension of Liquid Iron–Carbon Alloys. P. Kozakévitch, S. Chatel, and M. Sage. (*Compt. Rend.*, 1953, **236,** May 27, 2064–2066). The surface tension of iron–carbon alloys has been determined by the sessile drop method using an X-ray beam to reveal the shape of the drop within the furnace tube. The results can be represented by the equation $\gamma = 1600–100(C\ \%)$, applicable between 0·1 and 5% carbon. This demonstrates that the wetting relationships at the slag–metal interface are independent of the carbon content of the metallic phase.—A. G.

New Ultrasonic Test Method. T. A. Dickinson. (*Steel Processing*, 1953, **39,** July, 327–329). **Immersion Ultrasonic Inspection—The Development of a New Technique.** T. A. Dickinson. (*Welding and Metal Fab.*, 1953, **21,** Aug., 296–297, 308). The immersion ultrasonic inspection equipment developed by Electro-Circuits Inc., Pasadena, Calif., is described. Ultrasonic waves are produced and vibrate a quartz crystal at 5,000,000 or more times per second. These wave frequencies, when conforming with the resonance values of different inspection materials, are echoed or reflected back to their sources by non-resonant media, such as slag inclusions or blowholes within a weld. The new equipment uses a solution (usually water or oil) in which both quartz crystal and inspection material are immersed, therefore eliminating the necessity for an even surface to maintain close contact. The ultrasonic echoes are amplified and projected on to a screen of a special cathode tube, and a complete structural image is obtained.—U. E.

Ultrasonics in Industry. E. A. Neppiras. (*Research*, 1953, **6,** July, 271–282). A survey is made of the applications of ultrasonic techniques to underwater signalling ; non-destructive testing of solids; measurements of the elastic constants of metals ; fatigue testing ; machining of brittle materials ; dispersion, emulsifying, breaking up large molecules, and cleaning solids, in various liquids ; grain refinement, degassing and surface hardening of metals ; dust and smoke precipitation ; and processing in various chemical industries. Possible future applications are discussed. (35 references).—K. E. J.

Supersonic Inspection of Wire. A. E. Williams. (*Wire Prod.*, 1953, **2,** July, 4–7). The use of supersonic flaw detectors for the inspection of billets, rods, etc., for pipe, segregation, and hairline cracks is described. At present it is mainly used in the inspection of light metals. For steel wire and steel wire ropes, use of their magnetic properties is made in the 'Cyclograph', an instrument developed by the General Engineering Co. and the Nova Scotia Department of Mines ; this instrument and its operation are also briefly described.

Ultrasonic Inspection of Welded Pressure Vessels. (*Engineer*, 1953, **196,** Aug. 14, 199–202). This article describes the progress made in the field of ultrasonic testing techniques by W. J. Fraser and Co. Ltd., in devising and developing apparatus for inspecting plates and welded pressure vessels.

Non-destructive Testing Methods. C. C. Gee. (*Steam Eng.*, 1953, **22,** July, 377–381 ; Aug., 407–411). The various types of non-destructive test are set out together with their applications, and a brief discussion is given of the categories into which they are divided. The first part concludes with a short consideration of proof loading tests in which the article to be tested is stressed to a predetermined level above that at which it will have to operate in service. In the second part methods of detecting surface flaws are reviewed.—B. C. W.

Non-Destructive Testing. R. C. McMaster. (*Proc. Amer. Soc. Test. Mat.*, 1952, **52,** 617–688). The author discusses in detail methods used in the non-destructive testing of materials. A large number of different types of commercial equipment are described and illustrated. (161 references).—B. G. B.

Aspects of Nuclear Fission of Interest to Foundrymen and Metallurgists. E. W. Colbeck. (*Metallurgia*, 1953, **48,** July, 33–40). The application of radiographic testing methods for the examination of foundry castings is described. Radioactive materials have been used to detect cracks in castings and examples of their use for the investigation of foundry problems are discussed.—B. G. B.

Handling Radioactive Isotopes in Industry. L. Walter. (*Mech. World*, 1953, **133,** Sept., 396–398). Some industrial uses of radioactive isotopes are given. Precautions in handling them are discussed.—D. H.

Notes on the Definition of Radiographic Sensitivity. M. Robba. (*Alluminio*, 1953, **22,** Jan., 28–36). [In Italian]. The author examines the geometrical relations which must be satisfied in X-ray photography to get radiographic sensitivity. Conditions are established for the distances between source, defect and film to obtain satisfactory definition.—M. D. J. B.

Radiography and Cobalt-60. G. J. Barker. (*Amer.. Foundryman*, 1953, **24,** Aug., 44–46). A very brief review is given of the use of $cobalt_{60}$ in the radiography of steel castings. The health hazards arising from this technique are considered in detail.—B. C. W.

Process Control in Gun Tube Manufacture. H. C. Dill and F. B. Stern, jun. (*Metal Progress*, 1953, **63,** June, 161–162). The gun barrel is magnetized in such directions that cracks or discontinuities produce local magnetic leakage fields. When a suspension of ferromagnetic particles coated with a fluorescent material is applied to the magnetized gun tube they are attracted to these local magnetic fields and the outline of the crack or discontinuity can be observed when the tube is illuminated by ultra-violet light. The method has been used successfully for 18 months for examining the inside and outside of gun barrels.—B. G. B.

On the Invar Problem. Y. Tino. (*J. Sci. Res. Inst.*, 1952, **46,** Sept., 141–146). [In English]. This is an account of previous work and recent research. The low coeff. of expansion of iron–nickel alloys in the region of 36% Ni (" Invar ") is explained in terms of phase composition and demagnetization effects near the Curie point.—K. E. J.

A Rationalization of Measured High-Temperature Properties of Fe–Cr–Co–Ni Alloys. J. D. Nisbet and W. R. Hibbard, jun. (*Trans. Amer. Inst. Min. Met. Eng.*, 1953, **197 :** *J. Met.*, 1953, **5,** Sept., 1149–1165). The authors have studied the high-temperature properties of binary, ternary, and quaternary Fe–Cr–Co–Ni alloys and have analysed the effects of single additions of incongruous elements, using a time-temperature parameter, in order : (*a*) To determine the rupture strength at temperatures up to 2000° F. ; (*b*) to characterize the different alloys as solid solution, allotropic transformation, precipitation hardening, or multiple phase alloys ; and (*c*) to emphasize the importance of such factors as relative melting points, relative atom sizes, recrystallization and recovery, and their interactions. The results are presented graphically and the principles deduced are discussed.—G. F.

Creep and Rupture of Chromium–Nickel Austenitic Stainless Steels. E. J. Dulis, G. V. Smith, and E. G. Houston. (*Trans. Amer. Soc. Metals*, 1953, **45,** 42–76). The results obtained during the mechanical and creep testing of 18/8 Cr–Ni steels are presented in detail. A number of American steels in this class with and without the addition of other alloying elements have been examined. The changes in microstructure

and in certain room-temperature mechanical properties brought about by creep testing are described.—B. G. B.

Austenite Stability and Creep-Rupture Properties of 18/8 Stainless Steels. J. K. Y. Hum and N. J. Grant. (*Trans. Amer. Soc. Metals*, 1953, **45**, 105–133). Two series of simple unstabilized 18/8 Cr–Ni steels were tested, one a series with carbon from 0·001 to 0·18%, and the second a series with nitrogen from 0·005 to 0·176%. The effect of these composition changes on the structure and the effect of structure on the creep-rupture properties were then studied. Creep-rupture tests were run from 1000° to 1300° F. for rupture times of 3 sec. to about 1000 hr. Over these carbon and nitrogen ranges large differences in rupture strength and creep resistance are noted, the rupture strength going through a minimum and the creep rate through a maximum. The effects of carbon and nitrogen are compared, these assist in explaining the strength theory proposed. In these unstabilized steels, the amount of carbide or nitride is shown to be the most important strengthening effect in the usual carbon and nitrogen ranges.—B. G. B.

The Role of Anelasticity in Creep, Tension, and Relaxation Behaviour. J. D. Lubahn. (*Trans. Amer. Soc. Metals*, 1953, **45**, 787–838). The experiments described show that creep recovery exhibits the same characteristics as anelastic deformation. Both can be described in terms of the same mechanical model. In a Cr–Mo–V steel, the amount of creep recovery as a function of time is related to (*a*) the deviations from lineality of a stress-strain curve at small stress, (*b*) the temporary shortening and eventual lengthening following partial unloading. and (*c*) the stress relaxation behaviour (at constant length) following partial unloading. The observations are explained quantitatively by identifying creep recovery with anelastic deformation. There is a limiting stress for the Cr–Mo–V steel, below which creep is essentially all anelastic. This limiting stress depends somewhat on the duration of the creep test ; it is about 20,000 lb./sq. in. at 800° F. and can be raised to at least 40,000 lb./sq. in. by as little as 1% plastic strain at 800° F.—B. G. B.

Special Steels in Free Piston Gas Generator. (*Aciers Fins Spec. Franç.*, 1953, Mar., 98–100). A short description and the principles of operation of the free piston generator and gas turbine, with some details of their design and materials of construction, are given.—V. G.

Alloying Ferritic Steels for Creep Resistance. H. W. Kirkby. (*Alloy Metals Rev.*, 1952, **8**, Sept., 2–7). The creep-resistant ferritic steels and the influence of various alloying elements on creep properties are reviewed. The progress made towards the development of steels with improved creep resistance is discussed.—V. G.

Creep Resisting Ferritic Steels for Gas Turbines. H. W. Kirkby. (*Alloy Metals Rev.*, 1952, **8**, Dec., 2–6). The properties of creep-resisting ferritic steels in relation to possible industrial gas turbine application are reviewed.—V. G.

Thermal Shock and Other Comparison Tests of Austenitic and Ferritic Steels for Main Stream Piping—A Summary Report. W. C. Stewart and W. G. Schreitz. (*Trans. Amer. Soc. Mech. Eng.*, 1953, **75**, Aug., 1051–1072). Thermal-shock test results on ferritic and austenitic steels are summarized. Thermal-shock tests on 6 in. pipe and valve assemblies in both 18/8 stainless steel and 2·25%-Cr–1%-Mo steels are described. Results include tensile properties at up to 1100° F. and stress-rupture, creep, and fatigue properties at 1000° and 1100° F.—D. H.

Jet Engine Hot Parts: Use of Ceramic Coated Low Alloys. J. V. Long. (Amer. Soc. Mech. Eng. : *Iron Steel*, 1953, **26**, Aug., 382–386). The need for conservation of scarce alloying materials used in high-temperature components of jet engines has led to experiments with lower-alloy components protected with ceramic coatings. Trials made at the Solar Aircraft Co. in California are discussed and a number of such components which have been successfully developed are described and illustrated.—G. F.

Effect of Titanium on Low Carbon 3% Nickel Steel. T. Saito. (*Tetsu to Hagane*, 1953, **39**, Jan., 37–44). [In Japanese]. Titanium additions in the range 0–1·0% were studied, for the preparation of Ni–Ti steel for NH_3 synthesis equipment. The major effects of the addition appeared after conversion of all carbon to TiC. Denitriding effects were strong. Macrostructure changed to small uniform dendrites ; microstructure showed removal of cementite and appearance of fine TiC. Grain size was satisfactory below 0·3%. The A_1 and A_3 points rose with titanium content. Maximum hardening by tempering was produced by 0·6% Ti. In the normalized condition, tensile strength decreased slightly, but shock resistance improved greatly. Optimum toughness was obtained at 0·3–0·4% Ti.—K. E. J.

Effect of Tungsten on the High-Carbon High-Chromium Die Steel. S. Koshiba and S. Nagashima. (*Tetsu to Hagane*, 1953, **39**, Feb., 119–122). [In Japanese]. Investigations of the effect of tungsten on the properties of die steels are reported. A steel containing 1·25–2·00% W was found to be better than non-tungsten steel for punching and drawing dies.

Effect of Arsenic on Impact Values of Carbon Steels. H. Sawamura and T. Mori. (*Tetsu to Hagane*, 1953, **39**, Apr., 418–426). [In Japanese]. Results are given for a range of arsenic contents between 0·0 and 1·0%. Charpy values for 0·05 and 0·25% C steels decreased with arsenic content to 500° C., but showed no change above this temperature. With higher-carbon steels, the temperature at which impact values increased rose with arsenic content. All showed high impact resistance above 750° C. Austenite and ferrite grain sizes were only influenced in dead soft steel. The hardness gradually increased with increase of arsenic content.—K. E. J.

The Use of Silicon in Special Steel. (*Acier Fins Spec. Franç.*, 1953, Mar., 25–32). Sources of silicon, its properties, methods of production, and application as an alloying element in cast iron and steel are outlined.—V. G.

Experiences with Boron Steels in Production. H. B. Knowlton. (*Metal Progress*, 1953, **63**, Apr., 67–74). The replacement of alloying elements in steels (particularly nickel and molybdenum) by boron in the past few years in America is outlined. Boron steel accounted for 10% of the American output of alloy steel for engineering purposes in 1952. Alloy steels for heat-treating containing 0·4 to 0·6% carbon have been satisfactorily replaced by boron steels having the equivalent hardening properties. The service performance of these steels has been satisfactory. Substitution of boron for all or a part of other alloying elements, replaces only the hardening properties of that element and does not reproduce the hot strength characteristics.—B. G. B.

American Applications of Boron and Other Low-Alloy Steels. H. B. Knowlton. (*J. Iron Steel Inst.*, 1954, **176**, Feb., 187–205). [This issue].

Properties of Basic-Bessemer Low-Carbon Rimming Steel after Water-Quenching from Rolling Temperature. W. Dick. (*Stahl u. Eisen*, 1953, **73**, July 16, 945–965). An investigation was carried out to determine whether the properties of basic-Bessemer low-carbon rimming steel could be improved by simply quenching with water immediately after rolling. The water was sprayed at high and low pressure on to the surface of bars, sections, and heavy strip. All operations were carried out under works conditions, and tensile, impact, bend, and fatigue tests were made on aged and un-aged specimens consisting of lengths of bar, and section, or pieces machined therefrom. The influence of local heating, pressing or turning was also examined. Although improvements in mechanical properties were often found, the water-quenched steel did not satisfactorily fulfil the requirements of uniformity, deformability, and resistance to embrittlement necessary for heavy structural steels.—J. P.

Ultra High-Strength Steels. A. A. Bibee. (*Metal Progress*, 1953, **63**, May, 95–96, 167). A short account of the S.A.E. Conference at Los Angeles in March, 1953, to discuss the metallurgy, processing, and design of steel parts heat-treated to an ultimate tensile strength of above 90 tons./sq. in., is given. The primary use of such parts is in aircraft construction.

Wear-Resisting Manganese Steel for Crushing Machines. A. Rolla. (*Ing. Mecc.*, 1953, **2**, Mar., 19–20). [In Italian]. Wear and impact resisting components for crushing machines are manufactured from Hadfield manganese steel (C 1%, Mn 13%). These components, which may be cast or forged, should be used only after heat-treatment by quenching in water at a temperature of 1050° C., thus ensuring maximum resistance to wear and impact. Failure or poor resistance can usually be traced to incorrect heat-treatment. A recent development in the chemical composition of the Standard Hadfield Steel is the introduction of 3–5% Ni which eliminates the need for heat-treatment.—M. D. J. B.

New Chromium Steels of Low Nickel Contents and Their Applications. A. Roos. (*Génie Civil*, 1953, **140**, Aug. 1, 290–292). The development in America of a new chromium

steel in which the nickel is largely replaced by manganese without impairing mechanical properties or quenchability is described. The composition of this 'T.R.C. Steel' is 15·23% Cr, 0·95% Ni, 17·59% Mn, 0·08% C, 0·38% Si, and 0·13% N_2.—A. G.

Nickel Restrictions Bring Use of New Stainless Steels. R. A. Lincoln. (*Iron Age*, 1953, **171,** May 14, 129–132). As a new substitute for 18/8 stainless steel an alloy steel with 15·5% Cr, 17% Mn, and 0·9% Ni has been produced. Corrosion, weldability, mechanical properties, and rate of work hardening compare favourably. It is suggested that for best properties substitutes should contain a minimum of 3·5% Ni; the maximum at present allowed in the U.S. is 1% Ni. A steel with 13% Cr and 18% Mn is completely austenitic when annealed, but corrosion resistance is too low except for limited applications. (13 references).—A. M. F.

The Stainless Steels. G. Guidi. (*Acciaio Inossid.*, 1953, **20,** Jan.-Feb., 3–12; Mar.-Apr., 27–35; May-June, 51–60; July-Aug., 75–91). [In Italian]. The author makes a comprehensive study of the mechanical, chemical and corrosion-resisting properties of stainless steels and examines the effect of surface finish and heat-treatment on their resistance to corrosion.—M. D. J. B.

New Stainless Alloy Fills Long Industry Need. N. F. Mott. (*Iron Age*, 1953, **181,** June 18, 149–153). A new hard, corrosion resistant and non-galling stainless steel alloy V2B developed by the Cooper Alloy Foundry Co. is described. The alloy is a hardenable 18/8 stainless steel containing copper, molybdenum, silicon, and a very small amount of beryllium and, after machining, it may be precipitation hardened by low-temperature heat-treatment. A carefully controlled analysis giving 40–50% ferrite allows 0·10–0·20% Be to effect the hardening. Tables are given showing the effects of the above elements on the hardness obtained. The alloy can be produced in the cast or rolled condition.—A. M. F.

Non-Magnetic Steels in the Construction of Electrical Equipment. A. Rolla. (*Ing. Mecc.*, 1953, **2,** Feb., 23–24). [In Italian]. Parts of electrical machines which are highly stressed, must be made of non-magnetic steels to ensure that the magnetic field is not disturbed. Nickel-manganese steels, with stable austenitic structure, are non-magnetic even after severe work hardening, but they have a low yield strength. These steels, however, can be employed after controlled strain hardening, which brings their yield strength up to the required limit.—M. D. J. B.

Stainless Steels of Austeno-Ferritic Structure. (*Acier Fins Spec. Franc.*, 1953, Mar., 65–69). The structure of 18/8 stainless steel, causes of its intergranular corrosion, and the protecting influence of additions of titanium and niobium are explained. The production of an austenitic-ferritic stainless steel resistant to intercrystalline corrosion by increasing the chromium content to 20% and additions of 2·5% molybdenum and 1·5% copper is briefly described, together with its structure and properties.—V. G.

81B40. R. N. Imhoff and J. W. Poynter. (*Metal Progress*, 1953, **64,** July, 65–71). The results of laboratory tests carried out on 81B40 American boron steel (C 0·38–0·43%, Ni 0·20–0·40%, Cr 0·35–0·55%, Mo 0·08–0·15%, B 0·0005% min.) are given and compared with those of 8740 and 4140 non-boron steels. (Cr–Mo and Ni–Cr–Mo steels respectively). The mechanical properties of all three steels of the same hardness level are similar. The properties of the boron steel in the normalized state are not as good as those of comparable low-alloy steels without boron. The 81B40 material is somewhat suspectible to temper brittleness. The results so far indicate that 81B40 will be a suitable substitute for 8740 and 4140 steels.—B. G. B.

Transformer Steels—Their Choice and Present State of Development. C. D. Mee. (*Elect. Rev.*, 1953, **153,** Sept. 11, 551–555). The factors influencing the choice of material for power transformers are discussed, they include availability, cost, and suitable metallurgical and magnetic properties. The mechanism of magnetic induction is outlined, and the present state of development and possibilities of improvement are indicated.

Cold Drawn Prestressing Wire. J. L. Bannister. (*Structural Eng.*, 1953, **31,** Aug., 203–218). The conditions in which cold drawn high-tensile steel wire can be supplied are described together with the effects of various treatments on the behaviour of wires under service conditions. Results of investigations are discussed concerning the relaxation of stress and stress-strain relationship under varying tensile stresses.—E. C.

The Use of Deformed Steel Bars in Reinforced Concrete. (*Structural Eng.*, 1953, **31,** Aug., 222–225). The use of deformed bars instead of plain bars in reinforced concrete is outlined. Deformed bars are defined and suitable steels are quoted. Details of design are specified together with dimensional requirements. Extracts from A.S.T.M. specifications, and bending and bond tests are appended.—E. C.

Correlation Between Tensile Strength and Dimensions for Grey Cast Iron. M. Itzel and E. O. Lissell. (*Gjuteriet*, 1953, **43,** Aug., 143–147). [In Swedish]. Proposed Swedish standards for grey cast iron are based on the tensile properties of a single test bar. Tests were made to determine the probable tensile strengths of castings having dimensions different from those of the test piece.—G. G. K.

Suggested Standards for Grey Cast Iron. (*Gjuteriet*, 1953, **43,** Aug., 148–150). [In Swedish]. Three classes of cast iron have been standardized, having minimum standard test bar tensile strengths of 15, 20 and 25 kg./sq. mm. with corresponding maximum Brinell hardnesses of 200, 225 and 250. Rules are given for testing strength and hardness.—G. G. K.

Modern Cast Irons—A Digest of Data on Currently Used Materials. K. J. A. Brookes and R. H. Warring. (*Mech. World*, 1953, **133,** Sept., 393–401). The authors review the methods used to increase the strength, wear resistance and corrosion resistance of cast irons. Useful tables are given classifying, and summarizing the specifications and tabulating the mechanical properties of cast irons in current use.—D. H.

The Influence of the Primary Structure of Grey Cast Irons on Their Moduli of Elasticity. Elizabeth Plenard. (*Compt. Rend.*, 1953, **237,** July 20, 260–262). Curves relating the modulus of elasticity to the average length of the graphite lamellæ (primary structure) and to the nature of the matrix (secondary structure) have been constructed. These indicate that only the primary structure has any influence—a conclusion supported by the fact that ferrite and pearlite free from graphite have the same value of Young's Modulus.—A. G.

The Structure and Mechanical Properties of Modified Cast Iron. G. I. Kiezkin. (*Foundry Production* (Moscow), 1952, (7), 17–22; *Met. u. Giesserei Techn.*, 1953, **3,** June, 220–223). Increases in strength and plastic properties of nodular iron treated with magnesium can only be achieved at the cost of involved technical processes, unfavourable casting characteristics and high prices. The author therefore advocates more extensive use of high quality modified cast irons with laminar graphite formation provided plasticity is not a major consideration and tensile strength is between 35 and 40 kg./sq. mm.—L. J. L.

A Research Engineer Looks at Metallurgy. C. G. A. Rosen. (*Metal Progress*, 1953, **63,** Apr., 93–96, 194–196). The effect on the development of diesel engines which advances in metallurgy have made possible are discussed.—B. G. B.

METALLOGRAPHY

Progress in Metallurgical Microscopy. B. W. Mott. (*Endeavour*, 1953, **12,** July, 154–161). Improvements in existing methods and new techniques of metallurgical microscopy are surveyed, including applications of phase contrast, polarized light, and interference methods, also the use of reflecting objectives for examination of hot metal surfaces. (41 references).—E. C.

Electron Microstructure of Bainite in Steel. (*Proc. Amer. Soc. Test. Mat.*, 1952, **52,** 543–572). The results of an investigation into the structure of bainite in steel are given and discussed in detail. The structure of lower and upper bainite has been determined and the changes in structure during the transition from lower to upper bainite are described. Twenty-five micrographs are reproduced to illustrate the different structures.—B. G. B.

Detection of Microcracks in Steel. W. L. Jensen and R. F. Campbell. (*Trans. Amer. Inst. Min. Met. Eng.*, 1953, **197;** *J. Met.*, 1953, **5,** Sept., 1222). The authors have detected microcracks in quenched 1·14% C steel after mechanically or electrolytically polishing without intermediate etching, proving that such cracks are present before etching and are not the result of the etching treatment.—G. F.

Micro-Autoradiography As an Aid in the Testing of Materials. J. Rasch. (*Z. Metallkunde*, 1953, **44,** July, 321–322). By covering the surface of an alloy specimen containing radioactive constituents with a sensitized emulsion, followed

by a suitable exposure and processing, the position and distribution of the radioactive constituents can be determined, using an optical microscope. In the case of multiple alloys, phase analyses may be carried out by a similar method. Examples and illustrations are given.—L. D. H.

Pre-Shadowed Replicas for Electron Metallography. W. L. Grube and S. R. Rouze. (*Proc. Amer. Soc. Test. Mat.*, 1952, **52**, 573–591). The preparation of replicas for electron microscope examination is considered. A new method is described whereby thin metal films can be stripped successfully from etched metal surfaces. A special coating of a surface active agent is applied to the specimen before the metal film is applied. The advantages of this new technique are described. The pre-shadowing technique is shown to improve the resolution of the replica considerably. This involves the depositing of a thin film of a dense metal at an oblique angle upon the replica surface.—B. G. B.

Electron Diffraction. A. E. Williams. (*Mech. World*, 1953, **133**, Sept., 390–394). The author discusses briefly the theory of electron diffraction. Some mention is made of its application to the examination electro-deposits, thin films, and other surface phenomena.—D. H.

Gamma-Ray Metallography. M. Ezran. (*Rev. Gén. Méc.*, 1953, **37**, Apr., 119–122). The use of gamma rays for the examination of the soundness of materials is considered. For sources of intensity comparable to radium, cobalt$_{60}$ and tantalum$_{182}$ are used; these have a half life of 5·3 years and 4 months respectively. For a less intense source, such as is suitable for the examination of thin sections, iridium$_{192}$ with a half life of 2½ months, is used. Examples of the application of this technique are given.—B. G. B.

The Measurement of X-Ray Line Breadths. T. R. Anantharaman and J. W. Christian. (*Brit. J. App. Phys.*, 1953, **4**, May, 155–56). Simple analytical methods for finding the position of the α_1 peak and the integral breadth of the α_1 component of a composite Debye-Scherrer powder line are described. The expressions are exact for symmetrical lines, and sufficiently accurate for very broad asymmetrical lines. The problem of correct location of the background level is also considered.—J. O. L.

Modifications of the Schulz Technique for the X-ray Determination of Preferred Orientation in Rolled Metal. M. L. Fuller and G. Vaux. (*Trans. Amer. Inst. Min. Met. Eng.*, 1953, **197**: *J. Met.*, 1953, **5**, Aug., 1038). The authors describe two modifications to the Schulz method of determining by X-ray reflection the preferred orientation of crystals in rolled metal. The first provides automatic scanning and recording of the reflections; the second permits application of the procedure to coarse-grained specimens and allows measurement of almost all orientations without resorting to the transmission technique.—G. F.

An X-ray Camera for Measuring Preferred Orientation in Wires. A. L. MacKay. (*J. Sci. Instruments*, 1953, **30**, July, 244–245). A camera is described from which the degree of orientation of each reflecting plane can be read directly from a specially prepared chart. Estimates of strain or grain size can be determined at the same time.—L. D. H.

New Results with X-Ray Back-Reflection Method. F. Regler. (*Radex Rundschau*, 1952, (4), Aug., 166–180; (5), Oct., 200–209). The back reflection method is discussed. The possible accuracy of measurements is considered and the effects of target spot size and shape are described. A wide range of fields of application of the method is surveyed including the examination of ferromagnetic materials, determination of coefficients of thermal expansion, investigation of grain-size problems, structure distortions, and measurement of thickness of surface films. Sources of error and limitations of the method are discussed.—E. C.

Electrolytic Polishing of Steel at Elevated Temperatures. Z. Wójcik. (*Prace Instytutow Ministerstwa Hutnictwa*, 1953, **5**, (3), 123–128). [In Polish]. A method of investigating supercooled austenite in steel has been developed. It entails the electrolytic polishing at temperature up to 350° C. of specimens previously quenched at supercritical rates from above the Ac_3 point. The electrolytes and the technique of electropolishing at elevated temperature are described. The advantages and new applications of this method for studying the transformation of supercooled austenite, determination of grain size, investigation of incubation period, and the bainitic transformation are pointed out with examples.—V. G.

Microstructure. C. S. Smith. (*Trans. Amer. Soc. Metals*, 1953, **45**, 533–575). The microstructure of metals is considered in detail. The mathematical and physical factors which control the structure are discussed and numerous microphotographs are used to support the various structural theories which are explained.—B. G. B.

The Structure of Martensite Crystals in Hardened Steel. M. P. Arbuzov, L. I. Lysak, and E. G. Nesterenko. (*Doklady Akademii Nauk, S.S.S.R.*, 1953, **90**, (3), 375–377). [In Russian]. X-ray photographs were made of electrolytically isolated martensite from a hardened 0·98%-carbon steel using iron radiation. It is concluded from measurements of intensity and line width that: (1) The size of the area of coherent dispersion is small ($2-3\times10^{-6}$ cm.); and (2) the mean value of the square of the displacement in the direction of the *c* axis is twice that in the direction of the *a* axis.—V. G.

Preyield Plastic and Anelastic Microstrain in Low-Carbon Steel. T. Vreeland, jun., D. S. Wood, and D. S. Clark. (*Acta Metallurgica*, 1953, **1**, July, 414–421). The behaviour of an annealed low-carbon steel subjected to rapidly applied constant stresses less than the static upper yield stress is investigated and the relationship between the stress and the equilibrium microstrain determined. The results are explained by the generation of dislocations whose motion is arrested by grain boundaries. The authors are able to determine the characteristic length of a dislocation, the value being in agreement with the concepts of the mosaic block structure.—A. D. H.

The Effect of Temperature on the Rolling Texture of Plastically Deformed Low Carbon Steel Strip. N. P. Goss. (*Trans. Amer. Soc. Metals*, 1953, **45**, 333–343). Samples of low-carbon steel strip were rolled under standard conditions at temperatures between 32° and 800° F. X-ray diffraction patterns were taken by the reflection method. As the temperature of rolling is lowered from the normal temperature of rolling, the degree of preferred orientation increases. As the temperature is raised from the normal temperature of rolling a new type of structure is developed. These structural changes are discussed.—B. G. B.

A Grain Boundary Phenomenon in the Hardened Structure of Ball Bearing Steels. T. Hisamichi. (*Tetsu to Hagane*, 1953, **39**, Apr., 412–417). [In Japanese]. In the microstructure of hardened high-carbon low-chromium steels of spheroidized structure, the dark network structure along grain boundaries revealed by picrate etching was examined. It was shown to be martensite, arising from heretogeneity of elements such as carbon, probably caused by preferential solution of small cementite particles, and it had a marked influence on mechanical strength. The heterogeneity during austenitizing is attributed to the chromium content.—K. E. J.

Researches on Heat-Resistant Austenitic Alloys with Low Contents of Scarce Metals. W. Siegfried and F. Eisermann. (*Rev. Mét.*, 1953, **50**, July, 485–494). The influence of precipitates and the effect of alloying elements on the recrystallization of iron are considered, and the relation between the recrystallization limit and heat resistance is discussed. A refractory alloy suitable for cast parts able to withstand temperatures of 650–750° C. has been developed from an iron base material with 25% Ni and 18% Cr. It has been found possible to replace 8% of the nickel by equal amounts of manganese and cobalt, and the optimum additions of such minor elements as tungsten, vanadium, molybdenum, and niobium–tantalum have been determined.—A. G.

Notes on Work Hardening and Recovery. P. A. Beck. (*Acta Metallurgica*, 1953, **1**, July, 422–425). From a review of recent work on deformation, the author suggests that at least three stages occur in the annealing of a cold worked metal. They are: (1) Recovery, (2) sub-boundary migration, and (3) grain boundary migration (recrystallization with local reorientation). Only a small part of the work hardening is released in recovery, but sub-grain growth in (2) leads to the release of the larger remainder. In practice, stage (3) often sets in prematurely and takes over the function of relieving work-hardening.—A. D. H.

Location Correlation of Outer Electrons in Crystal Chemistry. K. Schubert. (*Z. Metallkunde*, 1953, **44**, June, 254–259). The author discusses examples of crystal structure in which the valency electrons do not exhibit an exclusively lattice type of positional correlation, and are statistically independent of the inner electrons. The allotropic transformation β-Mn → α-Mn is used as an illustration.—L. D. H.

Recrystallization and Stored Energy. H. P. Leighly, H. L. Walker, and J. W. Marx. (*Trans. Amer. Inst. Min. Met. Eng.*, 1953, **197:** *J. Met.*, 1953, **5,** June, 809–812). The authors briefly review the literature on recrystallization phenomena, and from a consideration of elementary statistical mechanics, predict a relationship between recrystallized grain size and previous deformation. Reasonable agreement with experiment is obtained.—G. F.

Interaction between Crystal Slip and Grain Boundary Movement. D. McLean. (*Nature*, 1953, **172,** Aug. 15, 300–301). During creep, grain boundaries migrate in a direction perpendicular to themselves, the extent being a function of the applied stress. An explanation of this phenomenon is put forward involving interaction between dislocations moving through the lattice and grain boundaries tending to halt them.

Theory of Slip-Band Formation. J. C. Fisher, E. W. Hart, and R. H. Pry. (*Phys. Rev.*, 1953, **87,** Sept. 15, 958–961). The fine structure of slip bands on the surfaces of plastically deformed crystals is explained in terms of the Frank-Read model of dislocation-loop generation. Back-stress at an active source is sufficient to stop dynamic loop generation at the source ; slip therefore occurs in time-separated avalanches. Avalanches from other systems crossing a slip plane containing an active source produce the observed markings.—K. E. J.

Observations on the Crystallographic Habit of Martensite. R. F. Mehl and D. M. Van Winkle. (*Rev. Mét.*, 1953, **50,** July, 465–475). The atomic and crystallographic mechanism of martensite formation and the influence of temperature on the predominant face has been investigated for plain carbon, nickel–chromium, and manganese steels. It is concluded that the predominant faces do not depend solely on the carbon content. For the steels studied there were two predominant faces, (225) and (259), whose frequency of appearance depended on the mean temperature of transformation.—A. G.

Some Impressions of Laboratories in the United States and of the Metallurgical Congress at Philadelphia. (October, 1952). A. R. Weill. (*Rev. Mét.*, 1953, **50,** July, 476–484). Among new techniques discussed at the above congress were : The production of monocrystals of such metals as titanium, zirconium, and vanadium for semi-conductors ; methods of studying textures and deformation in metals ; the investigation of magnetic domains by the methods of colloidal powders and of rotatory polarization ; and analysis by nuclear radiation.—A. G.

The Preparation in Sheet Form of Large Single Crystals of Silicon–Iron of Predetermined Orientation for Magnetic Purposes. R. G. Martindale and D. A. Langford. (*Proc. Inst. Elect. Eng.*, 1953, **100,** Aug., 417–426). A description is given of the preparation of large single crystals of a 2%-silicon iron. Factors affecting recrystallization are discussed, and limitations in orientation possibilities are pointed out. Magnetic-measurement/torque curves and magnetization curves are reproduced and the results interpreted.—L. D. H.

An Optical Goniometer for the Examination of Long Metal Single Crystals. A. J. Goss. (*J. Sci. Instruments*, 1953, **30,** Aug., 283–285). The goniometer described is designed to measure the orientation along their whole length of single crystals up to 30 cm. long. Microscopic facets are developed in the surface by chemical etching or thermal evaporation in a vacuum ; from these reflections, spots are produced on the goniometer scale. Etching techniques are given.—L. D. H.

Neumann Bands in Pure Iron. A. Kelly. (*Proc. Phys. Soc.*, 1953, **A66,** Apr. 1, 403–405). An X-ray microbeam method was used to examine Neumann bands in a single crystal of pure iron (0·3% Mn, 0·05% S and 0·05% P). The lattice in the bands was twinned with respect to the matrix about a (112) pole.—K. E. J.

On the Theory of Internal Friction in Metals. B. V. Paranjape. (*Proc. Phys. Soc.*, 1953, **A66,** June 1, 572–575). Absorption and emission of vibrational quanta by conduction electrons in metals leads to a contribution to the internal friction per cycle independent of frequency. An expression is derived for this, and its limitations indicated that it requires knowledge of the temperature coefficient of resistivity at high temperatures.—K. E. J.

The Order-Disorder Transformation Viewed As a Classical Phase Change. F. N. Rhines and J. B. Newkirk. (*Trans. Amer. Soc. Metals*, 1953, **45,** 1029–1055). It is shown that order-disorder transformations proceed as normal phase changes in accordance with the Gibbs phase rule. Presumed evidence to the contrary is found to have been based upon observations made without the necessary condition of equilibrium. It is proposed that the disordered phase be defined as a crystalline state characterized by the absence of stable long-range order, and that the ordered phase be defined as a crystalline state characterized by the presence of some degree of stable long-range order.—B. G. B.

Calculation of an Internal Stress Accompanying the Phase Change of Solids. S. Fujita. (*J. Sci. Res. Inst.*, 1952, **46,** June, 53–58). [In English]. Assuming that grains are spherical, and that transformation starts from their centres simultaneously, an expression for the magnitude of the internal stress accompanying a single solid-phase change is derived; it incorporates the volume increase, bulk and rigidity moduli, and spherical radii.—K. E. J.

The Redistribution of Solute Atoms During the Solidification of Metals. W. A. Tiller, K. A. Jackson, J. W. Rutter, and B. Chalmers. (*Acta Metallurgica*, 1953, **1,** July, 428–437). A quantitative analysis is made of the redistribution of solute resulting from solidification of a binary solution without convection under transient and steady state conditions. The distribution depends on the rate of solidification, an abrupt increase which can result in a band of high solute concentration being formed in the solid ; the reverse occurs when the rate is decreased rapidly. The length of the constitutionally supercooled zone adjacent to a growing solid-liquid interface is calculated.—A. D. H.

The Theory of D_0 in the Arrhenius Equation for Self-Diffusion in Cubic Metals. A. D. LeClaire. (*Acta Metallurgica*, 1953, **1,** July, 435–447). The author shows that in the equation $D = D_0 \exp(-Q/RT)$, the experimentally observed values for D_0 and Q are consistent only with the supposition of a vacancy mechanism in face-centred-cubic metals and a ring mechanism in body-centred-cubic metals. Theoretical relations are worked out to enable values of D to be estimated.—A. D. H.

Short-Range Order in Substitutional Solid Solutions in Metals. G. Fournet. (*Acta Metallurgica*, 1953, **1,** July, 383–389). [In French]. It is shown that short range order always exists in substitutional solid solutions. A method is given for calculating the parameters of short range order for a simple model.—A. D. H.

The Bauschinger Effect in Some Face-Centred and Body-Centred Cubic Metals. R. L. Woolley. (*Phil. Mag.*, 1953, **44,** June, 597–618). Work-hardenable metals, after plastic deformation and unloading, become anisotropic and will deform plastically under numerically smaller compression stresses. The effect has been studied in copper, aluminium, lead, nickel, and iron after strains between 1 and 120%. An expression for strain is derived, and dependence of the effect on grain size, temperature and purity is discussed. Results are compared with previous results and theories, and a qualitative explanation is advanced. The behaviour of iron is influenced by twinning. (15 references).—K. E. J.

Diffusion and Oxidation of Solid Metals. C. E. Birchenall. (*Indust. Eng. Chem.*, 1953, **45,** May, 907–911). A review of the major contributions during 1952 covers general literature, mathematical treatment of diffusion, self-diffusion coefficients, surface diffusion, grain boundary diffusion, the Kirkendall effect, dimensional changes, and a range of work on oxidation of metals and alloys. Work is reported on diffusivities in iron and oxidation phenomena in iron and several of its alloys. (103 references).—K. E. J.

Diffusion Grain Growth in Iron Alloys. F. Erdmann-Jesnitzer, H. Schumann, and M. Beckert. (*Arch. Eisenhüttenwesen*, 1953, **24,** May–June, 215–228). The published results on the formation of columnar crystals in low carbon steels during decarburizing have been critically assessed and extended by a series of experiments. Columnar grains occur under conditions of carbon migration; they have been found in the interior of cast iron as a result of the decomposition of supersaturated γ-iron. The orientation of the grains with respect to the direction of diffusion of carbon has been determined and a hypothesis to account for their formation is advanced. In agreement with this, it is found that, in the decomposition of austenite to pearlite in eutectoid steel, the α-iron matrix has its (120) plane parallel to the cementite lamellæ. Columnar grain growth should not be regarded as a recrystallization-diffusion process but as one of 'diffusion grain formation'.—J. P.

Free Energy of Vaporization of Metals from 0° to 2000° C. J. W. Evans. (*Trans. Amer. Inst. Min. Met. Eng.*, 1953, **197**: *J. Met.*, 1953, **5**, May, Section I, 655–657). This paper presents graphically the free energies of vaporization of a number of metals of economic importance, which can be used in conjunction with other free energy diagrams. The use of the diagram is illustrated by an esample.—G. F.

A Study of the Properties of Liquid Iron–Chromium Alloys by the Electromotive Force Method. O. A. Esin and N. A. Vatolin. (*Izvestiya Akademii Nauk S.S.S.R., Otdelenie Tekhnicheskikh Nauk*, 1953, (8), 1137–1142). [In Russian]. Measurements of e.m.f. were made at 1460° C. in a concentration cell made up of two Fe–C–Cr alloys and a slag containing CaO, MgO, SiO_2 and CrO. The chromium concentration was varied between 1·08 and 27·5%, and all alloys were saturated with carbon. From the results obtained, free-energy changes during the transfer of 1 g.-atom of chromium from a more concentrated to a dilute solution, and the activity coefficients of iron and chromium were calculated. From the values and mode of variation of the activity coefficients of the alloys studied, a positive deviation from ideal behaviour was indicated. Similar measurements were also made for Fe–Cr–Si–C alloys (saturated with carbon) using as an electrolyte a slag containing CaO, MgO, and SiO_2. Silicon concentrations were varied from 3 to 60% with the iron/chromium ratio (in percent) near unity. The curve representing the dependence of e.m.f. on silicon concentration has two points of inflection indicating the existance in the liquid alloy of atomic configurations corresponding to the silicides (Fe,Cr)Si and $(Fe,Cr)Si_2$. These results are in agreement with the two inflection points on the solubility curve of carbon in ferrochromium.—V. G.

The Influence of Plastic Deformation Occurring During the Decomposition of a Solid Solution, on the Velocity of Growth of Nuclei of a New Phase. L. N. Aleksandrov and B. Ya. Lyubov. (*Doklady Akademii Nauk S.S.S.R.*, 1953, **91**, (3), 519–522). [In Russian]. A solution of the problem of the rate of growth of a spherical nucleus of a new phase in a supercooled solid solution is attempted. The influence of plastic deformation on the surface of the nucleus as a result of the transformation itself is taken into consideration.—V. G.

Vacuum Furnaces for Melting or Processing of Metals. G. J. Crites. (*Metal Progress*, 1953, **63**, May, 161–162, 164–166). Vacuum furnaces heated externally by resistance wires are used for melting materials of low melting point and for the degassing and annealing of metals. The maximum temperature used is about 1850° F. Induction vacuum furnaces are used chiefly for melting materials which melt above 1850° F ; the induction coil is placed inside the vacuum chamber. Electric arc vacuum furnaces are used to melt metals of very high melting point (*e.g.*, zirconium). Using a consummable electrode made of the metal to be melted is considered to be the ideal method for melting reactive metals. Descriptions of these three types of vacuum furnace are given.

A Simple High-Temperature Induction Furnace. I. Gordon. (*Amer. Ceram. Soc. Bull.*, 1953, **32**, June, 207–208). An easily constructed H.F. induction furnace for use in air is described. Temperatures attainable with the furnace appear to be of the order of 2500° C. The furnace is useful for determining melting points or for rapid high-temperature coating experiments.—J. R. P.

The Theory of Precipitation Processes in Supersaturated Solid Solutions. W. Wepner. (*Arch. Eisenhüttenwesen*, 1953, **24**, May–June, 275–279). The theoretical studies of the precipitation process from supersaturated solid solutions from the view point of migration and interchange of dissolved atoms are considered in relation to experimental findings. A model is proposed in which the dissolved atoms in dislocations form the nuclei for the precipitation of a new phase. These nuclei grow by deposition of further atoms. On the basis of this concept, a new precipitation equation is derived and it is shown that it agrees better with the experimental values of Wert than does the formula of Wert and Zener. The existence of a limiting temperature for the activity of dislocations and the conclusions to be drawn therefrom as to the variation of rate of precipitation with temperature are discussed.—J. P.

Concentration Gradients Associated with Growing Pearlite. R. E. Grace. (*Trans. Amer. Inst. Min. Met. Eng.*, 1953, **197**: *J. Met.*, 1953, **5**, June, 820–821). Carbon concentration gradients preceding an advancing pearlite interface have been studied by partly reacting an alloy to pearlite, quenching to stop the growth, then tempering to precipitate carbide particles in the martensitic areas. From electron microscope studies, it does not appear that gross composition changes occur in austenite more than a fraction of an inter-lamellar spacing ahead of the growing pearlite.—G. F.

Calculation of Martensite Nucleus Energy Using the Reaction-Path Model. J. C. Fisher and D. Turnbull. (*Trans. Amer. Inst. Min. Met. Eng.*, 1953, **197**: *J. Met.*, 1953, **5**, July, 921–922). The authors estimate the energy associated with the formation of a critical-size martensite nucleus, starting with the ' reaction-path ' model and including a surface free-energy barrier. A simple dependence of interfacial free energy on shear angle is assumed.—G. F.

Martensite Nucleation in Substitutional Iron Alloys. J. C. Fisher. (*Trans. Amer. Inst. Min. Met. Eng.*, 1953, **197**: *J. Met.*, 1953, **5**, July, 918–920). The author examines the predictions of nucleation theory with respect to martensite transformation in substitutional Fe–Ni alloys. The maximum nucleation rate, as a function of temperature, is measurable only in an extremely narrow composition range. Experimental observation, including in some alloys completely isothermal transformation and in others the composition-dependence of the temperature of the beginning of transformation, are compared with the calculated behaviour.—G. F.

Effect of Nitrogen on the Properties of Cementite. M. Bridelle and A. Michel. (*Rev. Mét.*, 1953, **50**, June, 410–412). Cementites containing nitrogen have been prepared by the action of a carbon-monoxide/hydrogen/ammonia mixture on ferric oxide or by the decomposition of a carbonitride. The effect of nitrogen on the curie point and the stability of cementite and austenite has been studied. A strong similarity in the behaviour of sulphur and nitrogen in the cementite lattice was observed—nitrogen having a weaker action than sulphur.—A. G.

How Cold Treatments Improve Performance of Materials. J. L. Everhart. (*Mat. Methods*, 1953, **37**, Feb., 115–118). The principle applications of refrigeration in the materials field are discussed. The author deals with dimensional stabilization by sub-zero transformation of austenite, the prevention of ageing of aluminium alloys, the use of liquid carbon dioxide as a coolant in machining and grinding operations, and shrink fitting.—P. M. C.

Effect of Nickel and Molybdenum on Stabilization of the Austenite-Martensite Transformation. D. J. Blickwede. (*Trans. Amer. Inst. Min. Met. Eng.*, 1953, **197**: *J. Met.*, 1953, **5**, July, 922–923). The author has studied the effects of nickel and molybdenum on austenite stabilization in hyper- and hypo-eutectoid steels. Nickel below 1% has little effect, but above this content stabilization is increased at low carbon contents and decreased at high carbon contents. This effect is discussed on the basis of the reaction-path theory. Molybdenum markedly increases stabilization at high-carbon levels.

Reactions of Solid Iron with Melts of Aluminium and Aluminium Alloys. E. Gebhardt and W. Obrowski. (*Z. Metallkunde*, 1953, **44**, Apr., 154–160). A systematic examination is described of the layers built up in solid iron immersed in baths of aluminium, both pure and alloyed, with a large number of metals, and silicon. The rate of growth of the coating was noticeably altered by small alloying additions, and, in particular, silicon restricted the layer formation appreciably. Zinc added in quantities greater than 1% had a very deleterious effect.—L. D. H.

An Interpretation of Hysteresis Loops in A_3 and A_4 Transformations of Pure Iron. K. Honda and M. Sato. (*J. Sci. Res. Inst.*, 1952, **46**, Mar., 1–7). [In English]. Transformation of an iron cylindrical specimen is assumed to proceed radially inwards. Large internal stresses cause the hysteresis loops in the A_3 and A_4 transformations. Ac_3 and Ar_4 points will be brought together by tension, and Ar_3 and Ac_4 points separated by pressure. These four points may be considered as points on the dilatometric curve where the whole mass has just transformed on heating on cooling.—K. E. J.

Research on Annealing of the White Cast Iron. II. Z. Ueda. (*Tetsu to Hagane*, 1953, **39**, Feb., 129–134). [In Japanese]. Two trends in the behaviour of cementite during annealing are described. With low heating rates, the rate of disappearance of eutectic cementite decreases with increase of heating rate in the $Ac_1 - AC_3$ transformation range, mainly as a result of diffusion. Above a certain heating rate, however, the rate of disappearance increases with heating rate. This is

attributed to superheating and the stress accompanying transformation.—K. E. J.

Influence of Alloying Elements on the M_s Temperature of High-Speed Steel. M. Okamoto and R. Odaka. (*Tetsu to Hagane*, 1953, **39**, Apr., 426–432). [In Japanese]. High-speed steels of the 18/4/1 W–Cr–V type were studied. The M_s temperature was lowered by carbon, tungsten, chromium and molybdenum, was independent of aluminium addition, and was raised by vanadium, titanium, and boron : these results are attributed to carbide effects. With cobalt, it was raised by adding up to 6%, then lowered : this effect is related to austenite solubility changes. (12 references).—K. E. J.

Reaction Velocity of Metallic Phases in the Range of Allotropic Transformations. F. Erdmann-Jesnitzer and L. Marx. (*Metall*, 1953, **7**, July, 499–501). Measurements are described of the rate of dissolution of iron in the range of the α–γ transformation in aluminium. The experiments were performed to test the Hedvall rule concerning the appearance of a maximum reaction rate in the range of temperatures of allotropic change. The presence of discontinuities in the curve of reaction rate v. temperature in this range, as observed by Fischbeck, was confirmed. Hedvall's rule did not apply.

Isothermal Temper Embrittlement of SAE 3140 Steel. F. L. Carr, M. Goldman, L. D. Jaffe, and D. C. Buffum. (*Trans. Amer. Inst. Min. Met. Eng.*, 1953, **197**: *J. Met.*, 1953, **5**, Aug., 998). To assist in elucidating the mechanism of the temper brittleness reaction, the portion near 525° C. of the time-temperature transformation diagram for isothermal temper embrittlement of SAE 3140 steel (1·25% Ni, 0·60% Cr) has been redetermined. The results are shown graphically.—G. F.

Precipitation of Cementite from Solid Solution and Precipitation Hardening of Mild Steel. Z. Wusatowski. (*Hutnik*, 1953, **20**, (7), 229–233). [In Polish]. Changes in the mechanical properties of mild steel due to the precipitation of oxides and cementite from solid solutions are discussed.

The Martensite Transformation Temperature in Titanium Binary Alloys. P. Duwez. (*Trans. Amer. Soc. Metals*, 1953, **45**, 934–940). The temperature at which the martensite transformation from β solid solution to α prime supersaturated solution takes place has been measured in binary alloys of titanium with niobium, tantalum, tungsten, iron, chromium, and manganese. In all cases the M_s curve decreases with increasing amounts of all elements.—B. G. B.

The Kinetics of Martensite Formation. E. Scheil. (*Härterei-Technische Mitteilungen*, 1950, **5**, 9–43). The constitution and mode of formation of martensite are discussed, with references to the literature. Martensite formation is compared and contrasted with other transformations. (79 references).—L. D. H.

Transformation in the Intermediate Stage. A. Rose. (*Härterei-Technische Mitteilungen*, 1950, **5**, 44–64). The microstructures obtained in the transformation of (*a*) a 4·75% chromium steel, (*b*) a 1·17% chromium 0·15% vanadium steel, (*c*) a 1·15% manganese steel, and (*d*) two vanadium steels, have been studied experimentally and theories governing the reactions in the intermediate transformation stage are advanced.—R. A. R.

Formation of Ferrite in Hypo-Eutectoid Plain Carbon Steels. C. Margaretha Hickley and J. H. Woodhead. (*J. Iron Steel Inst.*, 1954, **176**, Feb., 129–139). [This issue].

Carbides Formed During the Tempering of Martensite. J. Pomey and R. Coudray. (*Compt. Rend.*, 1953, **237**, July 6, 62–64). Carbide formation during the tempering of martensite has been studied by differential thermomagnetic analysis. Results indicate that the hexagonal percarbide ξ appears first, but being very unstable, it transforms to the rhomboidal ζ form. Finally the latter form changes to cementite.

Rate of Propagation of Martensite. R. F. Bunshah and R. F. Mehl. (*Trans. Amer. Inst. Min. Met. Eng.*, 1953, **197**: *J. Met.*, 1953, **5**, Sept., 1251–1258). The authors describe a rapid amplifier technique capable of measuring accurately the very short durations of such phenomena as martensite formation, slip, twinning, and cleavage. By this method the time of formation of one martensite plate 0·01 in. long in a Fe–29·5%-Ni alloy has been measured as 3×10^{-7} sec. and its rate of propagation as approx. 3300 ft./sec.—G. F.

Kinetics of First-Stage Graphitization in Iron–Carbon–Silicon Alloys. J. Burke and W. S. Owen. (*J. Iron Steel Inst.*, 1954, **176**, Feb., 147–155). [This issue].

Thermodynamic Activities in Iron–Nickel Alloys. R. A. Oriani. (*Acta Metallurgica*, 1953, **1**, July, 448–454). The activities of Fe–Ni alloys are studied in a differential apparatus by equilibrating H_2O/H_2 mixtures over the alloys and pure iron. A correction is made for the variation in activity of wüstite with ambient gas composition, but the formation of magnetite prevents results being obtained for nickel-rich alloys. Iron-rich face-centred-cubic alloys are shown to form nearly ideal solid solutions.—A. D. H.

On a Non-Magnetic State Found in the Irreversible Fe-Ni Alloys. Y. Tino. (*J. Sci. Res. Inst.*, 1953, **47**, Mar., 7–11). [In English]. Large thermal hysteresis divides the iron–nickel alloys into ' irreversible ' and ' reversible ' groups, each showing mixtures of α and γ phases subjected to local distortion. The border zone is the in region of 30% Ni. The fact that these mixtures, not in equilibrium, are in a stable condition leads to various anomalous properties in the alloy systems.—K. E. J.

A New Exposition of the Fe-Ni Phase Diagram. Y. Tino. (*J. Sci. Res. Inst.*, 1952, **46**, June, 47–52). [In English]. Two general versions of the phase diagram for the solid portion of the iron–nickel system have appeared, based respectively on thermal and magnetic analyses, and X-ray analysis. The discrepancies are attributed to the impossibility of supercooling phases stable at high temperature down to room temperature, and to high thermal hysteresis. A reconciliation of the two types of diagram is put forward, based mainly on free energies of α and γ phase.—K. E. J.

CORROSION

Corrosion of Iron and Steel and Its Prevention. H. Steinrath. (*Stahl u. Eisen*, 1953, **73**, Aug. 13, 1109–1116 ; Sept. 24, 1302–1308). The literature on corrosion of iron and steel and its prevention, published during 1950 and 1951, is reviewed. (397 references).—J. P.

The Biological Corrosion of Iron. G. Seelmeyer. (*Werkstoffe u. Korrosion*, 1953, **4**, July, 241–247). This is a general article about the various types of micro-organisms that can cause, directly or indirectly, the corrosion of iron. The author describes the individual organisms, discusses the ways in which microbiological corrosion can manifest itself, and outlines possible preventive measures —J. C. H.

Experimental Evidence of the Electrochemical Mechanism of Corrosion. U. R. Evans. (*Proc. 11th International Congress of Pure and Applied Chemistry*, 1947, **5**, July, 743–758). The electric currents flowing over metallic surfaces corroding in salt solutions have been measured by six different methods. The current strength is always of the right order of magnitude to account for the corrosion occurring and in some cases is exactly that calculated from Faraday's Law. A seventh method of measurement shows that corrosion only occurs at places which are anodic. During corrosion by acid solutions electric currents have been detected flowing between redeposited impurities (cathodes) and the parent metal (anode). Dry oxidation and tarnishing also depend on the electric transport of ions and electrons moving through the film. This is shown by the agreement between the observed rate of growth and the values calculated from the electrical properties of the film and the help of Wagner's equations.—B. G. B.

Corrosion—Its Cause and Control. W. R. Thomas. (*Canad. Metals*, 1953, **16**, June, 21–22).

Effect of Rust and Environment on Inhibition by Zinc-Rich Paints. J. E. O. Mayne. (*J. Iron Steel Inst.*, 1954, **176**, Feb., 140–143). [This issue].

Atmospheric Exposure Tests with Zinc-Rich Polystyrene Paints. J. E. O. Mayne. (*J. Iron Steel Inst.*, 1954, **176**, Feb., 143–146). [This issue].

Rust Protection for Machined and Finished Parts. (*Canad. Metals*, 1953, **16**, June, 22–24).

Magnetic Amplifier Controls for Rectifier Protecting Underground Metallic Structures Cathodically. O. Henderson. (*Corrosion*, 1953, **9**, July, 216–220). Circuit and characteristics of a magnetic amplifier control are described. It is preferred to other types of rectifier controllers.—J. F. S.

Protection of Metals Against Corrosion. F. Ritter. (*Österr. Masch. Elektrowirtschaft*, 1953, **8**, July, 202–204). After dealing briefly with the various types of metallic corrosion, the author summarizes the chief measures adopted to give protection. Subjects treated include the use of inhibitions in pickling, and oraganic and inorganic protective layers.—L. D. H.

Cathodic Protection of Iron and Steel against Corrosion. H. Klas. (*Stahl u. Eisen*, 1953, **73**, July 16, 971–974). The conditions for effective cathodic protection, particularly of buried steel pipes, are reviewed and procedures for determining the necessary values of the protective current are given. In this connection the potential difference between anode and cathode, the surface condition of the pipes and the electrical conductivity of the electrolyte (soil) are all important. For applying a protective current, iron or steel scrap or graphite is used for the anodes; only zinc, magnesium, aluminium or their alloys enter into consideration for sacrificial anodes. The strength of the current required depends on the extent to which the pipes become covered with an adherent deposit during service.—J. P.

The Elements of Cathodic Protection. K. A. Spencer. (*Inst. Petroleum Rev.*, 1953, **7**, May, 149–154). A short review is made of methods employed for the cathodic protection of metals. There is a balance between the cost of coating the metal and the cost of cathodic protection to obtain the optimum protection and this varies according to site and soil conditions. The economics of cathodic protection are discussed. For a 12-in. oil pipeline across Scotland the cost of cathodic protection is of the order of £3 per mile per annum.

Cathodic Protection Applied to Tank Bottoms. S. P. Ewing and J. S. Hutchison. (*Corrosion*, 1953, **9**, July, 221–231). A scale model for calculating current density requirements across cathodically protected tank bottoms is described. The theoretical considerations are discussed, and initial comparisons with actual tanks are reported.—J. F. S.

A Method of Evaluating Corrosion by Molten Metals. M. A. Cordovi. (*Proc. Amer. Soc. Test. Mat.*, 1952, **52**, 1027–1033). The method described has been used to investigate the corrosion by molten bismuth and bismuth alloys of a number of materials including several grades of steel. Two types of sample are used and are sectioned and photographed after treatment. The first consists of a precision ground solid cylindrical core which is cold pressed on to a machined hollow cylinder of the same material. After treatment in the corrosive molten metal the core is removed and the unaffected interior surface of the hollow cylinder provides a reference circle from which the maximum and average attack may be accurately measured. The second is in the form of a hollow cylinder which allows the corrosive molten metal to flow through and round the sample.—B. G. B.

Corrosion. M. G. Fontana. (*Indust. Eng. Chem.*, 1953, **45**, May, 91A–94A). A chart describes the corrosion resistance of a 14·5%-Si iron ('Duriron') to nitric acid at all normal concentrations and temperatures. It has excellent resistance to 45% and stronger acids at all temperatures including boiling; it is good at all concentrations below 16C° F. The alloy 'Durichlor' (containing approx. 3% Mo) is no more resistant. Chemical, physical, and mechanical properties of high-silicon irons, and their uses for industrial plant, are described.

A Laboratory Method for Screening Oil Well Corrosion Inhibitors. J. A. Caldwell and M. L. Lytle. (*Corrosion*, 1953, **9**, May, 186–187). A description is given of a laboratory test which has been used for the preliminary selection of inhibitors for use in oil wells. The test relies on weight loss measurements on steel panels exposed to the inhibited oil. Correlation with field tests has been good.—J. F. S.

Behaviour of Cast Iron in Sea Water; An Annotated Bibliography. Part I—Corrosion. Part 2—Test Procedures. (*Bull. Found. Abs. Brit. Cast Iron Res. Assoc.*, 1953, **12**, July, 8–14).

A Case History of Corrosion. C. R. Davis. (*Amer. Gas Assoc., Monthly*, 1953, **35**, May, 23–25). The methods employed for the protection of underground gas pipes owned by the Montana Power Company in America are described. Corrosion occurred in districts of widely differing soil types and the reasons for the particular method chosen for each situation are explained. Difficulties due to the proximity of other underground systems have been successfully overcome.

Examination of 335 Miles of Asphalt Mastic Coated Pipe. D. E. Miltner. (*Corrosion*, 1953, **9**, July, 210–215). Only one discontinuity per four miles was found by an electrical survey. Two-thirds of excavated breaks were due to friable mastic. The cathodic protection applied to the pipe had protected it well for nine years.—J. F. S.

Effect of Composition of Steel on the Performance of Organic Coatings in Atmospheric Exposure. F. L. LaQue and J. A. Boylan. (*Corrosion*, 1953, **9**, July, 237–241). Especially in corrosive atmospheres, paint performed better on low alloy steels than on open-hearth iron. A further improvement in corrosion resistance resulted from a phosphate treatment before painting.—J. F. S.

Corrosion Cracking of Martensitic Stainless Steel. A. E. Durkin. (*Metal Progress*, 1953, **64**, July, 72–75). The cracking of 12%-Cr stainless steel is shown to be the result of hydrogen embrittlement, not stress corrosion. The ductility of embrittled steels is shown to be recovered by immersion in hot water.—B. G. B.

Corrosion. M. G. Fontana. (*Indust. Eng. Chem.*, 1953, **45**, July, 91A–94A). Iso-corrosion curves for corrosion rates of 5, 20, and 50 mils/year for Durimet 20 and Carpenter 20 by nitric acid in relation to temperature are presented. Both these alloys are austenitic stainless steels.—R. A. R.

On the Black Residue in Treating Carbon Steel by Warm H_2SO_4 Method. II. A. Kiuchi. (*Tetsu to Hagane*, 1953, **39**, Feb., 123–128). [In Japanese]. Investigations were made on piano and rope wire rods of different geographical origin. Factors influencing the production of residue are discussed. Some residue analyses are given.—K. E. J.

The Use of Artificial Radioactive Isotopes for the Study of Diffusion and Self-Diffusion in Alloys. Self-Diffusion in γ Iron. P. L. Gruzin. (*Izvestiya Akademii Nauk S.S.S.R., Otdelenie Tekhnicheskikh Nauk*, 1953, (3), 383–392). [In Russian]. A method of determining coefficients of diffusion and self-diffusion, based on the use of radioactive isotopes, which can be used for any solid suitable for mechanical working is described. More exact results were obtained when using γ-radiating isotopes. The temperature dependence of the coefficient of self-diffusion of γ iron in the 960–1250° C. range was determined.—V. G.

Intergranular Corrosion of Nitrided Steel in Sulphuric Acid Containing Additions. S. A. Balezin and V. B. Ratinov. (*Doklady Akademii Nauk S.S.S.R.*, 1953, **90**, (3), 403–404). [In Russian]. Nitrided steel suffers intergranular corrosion only in sulphuric acid of concentration $\leqslant 0{\cdot}3N$. However, the use of a more concentrated sulphuric acid ($0{\cdot}5$–$4{\cdot}0N$), with additions effectively inhibiting the velocity of solution of the whole surface of the nitrided steel, causes local corrosion at boundaries of polyphase contact due to the presence of a dielectric on the surface of the steel. The depths of corrosion in the nitrided layer are tabulated. Non-nitrided steel also suffers similar corrosion, but its depth is much less. The corrosion phenomena is explained by the disturbance of the equipotential of the metal surface at boundaries of a polyphase contact and the adsorption at the metal acid-dielectric boundary of gaseous hydrogen which forms a fourth phase.—V. G.

Corrosion Tests on Metallic Materials. A. Ferri. (*Ing. Mecc.*, 1953, **2**, Jan., 19–23; Feb., 5–12). [In Italian]. The author sets out the objectives to be attained by corrosion tests on metallic materials. The principles on which these tests are based are discussed and descriptions are given of many laboratory and field corrosion tests. (23 references).

Atmospheric Corrosion of Steel Wires. A. P. Jahn. (*Proc. Amer. Soc. Test. Mat.*, 1952, **52**, 987–1004). The results of an atmospheric exposure test of wire and wire products which is now in its fifteenth year are summarized. The wires are mostly of galvanised steel. The data are discussed, first with respect to the performance of zinc coatings and secondly with respect to the performance of bare wires from which the zinc has been lost.—B. G. B.

Rust Protection by Means of Cement. H. Hebberling. (*Werkstoffe u. Korrosion*, 1953, **4**, July, 247–248). The author considers that the protective value of a cement or concrete coating on steel depends more on its pH value (alkalinity) than on the depth of cover. To ensure this alkalinity, concrete used to encase steel in structural work should contain at least 300 kg. of cement per cu.m. (500 lb./cu. yd.). Thin cement slurry is not always a satisfactory temporary protective coating for steel reinforcements. The application of a cheap grade of redlead paint is recommended; this paint should be removed by "flame-cleaning" before the steel framework is encased in concrete.—J. C. H.

Some Corrosion Problems Encountered in Mina-Al-Ahmadi Refinery. V. J. Chir, F. H. Colenutt, and K. Turner. (*Inst. Petroleum Rev.*, 1953, **7**, May, 159–164). Two different types of corrosion problem which have been encountered in the Kuwait refinery are discussed. The first is corrosion of the plant during the actual processing of the crude oil and the second is due to the use of sea water for cooling purposes. Modifications of the plant operation, of the constructional

materials used, and of the use of magnesium anodes have reduced the corrosion occurring.—B. G. B.

Resistance of Fe–Ni–Cr Alloys to Corrosion in Oxidizing and Reducing Flue-Gas Atmospheres. J. H. Jackson, C. J. Slunder, O. E. Harder, and J. T. Gow. (*Trans. Amer. Soc. Mech. Eng.*, 1953, **75,** Aug., 1021–1035). Corrosion of Fe–Ni–Cr castings by sulphur-bearing gases at 1800–2000° F. has been investigated. Sulphur varied from 0 to 500 grains/100 cu. ft. in both oxidizing and reducing flue gases. A wide range of compositions of the ternary system Fe–Cr–Ni was studied, the chromium varying from 11 to 36% with nickel up to 70%. In general the 28% Cr, 9% Ni, the 26% Cr., 20% Ni and the 26%-Cr 12%-Ni types had the best resistance under a variety of conditions. Corrosion in the higher sulphur-bearing atmospheres was much less severe when the flue gas was oxidizing than when reducing.—D. H.

Laboratory Investigation of Superheater Tubing Materials in Contact with Synthetic Combustion Atmospheres at 1350° F. C. J. Slunder. A, M. Hall, and J. H. Jackson. (*Trans. Amer. Soc. Mech. Eng.*, 1953, **75,** Aug., 1015–1019). A group of commercially available alloys including Inconel and various Ni–Cr austenitic stainless steels were tested for resistance to synthetic combustion atmospheres at 1350° F. to determine their suitability for superheater tubing materials. Two concentration levels of SO_2, CO, and alkali in the fuel were tested in all combinations. Evaluated on the basis of scaling and sub-surface corrosion, the 25/12 and 25/20 Cr/Ni types showed the best combination of properties.—D. H.

Corrosion of Mercury Boiler Tubes During Combustion of a Heavy Residue Oil. A. M. Hall, D. Douglass, and J. H. Jackson. (*Trans. Amer. Soc. Mech., Eng.*, 1953, **75,** Aug., 1037–1049). The authors discuss experiments carried out in an attempt to reduce tube wastage in a power-station mercury boiler. Sicromo 5S (5% Cr, 1·5% Si 0·5% Mo) steel tubes were coated with various sprayed metallic and refractory layers and subjected to a corrosion test under simulated service conditions. The use of ceramic coatings was discontinued because of brittleness and poor adhesion of the metallic layers; an undercoat of aluminium followed by a 20% Cr–Fe alloy showed much promise. An addition of calcium soap derived from tall oil when added in the ratio calcium atoms in the tallate.—D. H.

ANALYSIS

Stable, Sensitive Linear Starch Indicator Solution for Iodimetry. J. L. Lambert. (*Analy. Chem.*, 1953, **25,** June, 984–985). The preparation of a stable solution of linear potato starch fraction in 10% acetic acid is described. This is a very sensitive end-point indicator and 1–2 mls. are sufficient in place of the normal 3–5 mls. in a volume of about 200 ml.

Determination of Tin. Methods of Analysis Committee. (*J. Iron Steel Inst.*, 1954, **176,** Feb., 156–157). [This issue].

Several Points for Practical Analysis in Determining Manganese in Iron and Steel. (Gakushin Method). T. Iwahashi and Z. Yamaguchi. (*Tetsu to Hagane*, 1953, **39,** Feb., 140–148). [In Japanese]. As a result of discrepancies between the Gakushi (ammonium-persulphate/sodium-arsenite) laboratory method and a colorimetric method of estimating manganese in steel, was developed. Some points which might cause errors in the former method are clarified.—K. E. J.

Electrical Control of Spectrochemical Analysis. H. Gantzckow. (*Metall*, 1953, **7,** July, 517–519). The problem of electrical control of spectrochemical apparatus is discussed, and means of controlling and monitoring the discharge equipment are described.—J. G. W.

Improvement in Analytical Method of Hydrogen in Iron and Steel. Y. Yasuda. (*Tetsu to Hagane*, 1953, **39,** Apr., 432–436). [In Japanese]. The Gakushin process for determining hydrogen was examined. Suggestions are put forward for improving steel sampling procedures, and pretreatment and analysis of the cast samples.—K. E. J.

Improved Suction Apparatus for Sampling Liquid Steel for Hydrogen Determination. K. G. Schmitz and F. Willems. (*Stahl u. Eisen*, 1953, **73,** July 30, 1057–1058). The method for the determination of hydrogen in steel depends on taking a sample in a quartz pipette, quenching it rapidly in mercury and holding it at room temperature till all diffusible gas is evolved. This is analysed for hydrogen and the amount so found is added to that determined by fusing the sample from the pipette with tin. The apparatus described is for taking and storing the samples in such a way that no gas is lost and so that the hydrogen evolved is not diluted by inert gas.—J. P.

ECONOMICS AND STATISTICS

Metal Canada: 1952 Review. H. McLeod. (*Canad. Metals*, 1953, **16,** May 1, 6–22; May 20, 6–22). This is a complete statistical review of the Canadian metal industry for the year 1952. It was a peak year in steel production and automobile output. These figures are compared with those for previous years and they show Canada's position in comparison with the rest of the world.—J. C. B.

Iron and Steel Making in Mexico: Achievement and Potential. (*Times Rev. Indust.*, 1953, **7,** Sept., 99–100). Of the 303,400 tons of pig iron produced in Mexico in 1952, 300,000 tons were made at the Monterrey and Monclova plants. The three principal steelworks at Monterey, Monclova, and Piedras Negras produced 535,000 tons of ingots.—R. A. R.

Volta Redonda Produces More Steel. (*Eng. Min. e Met.*, 1953, **18,** May-June, 199). [In Portuguese]. In 1954 the Volta Redonda integrated iron and steel plant will produce 710,000 ingot tons. In 1952, the outputs of coke and basic pig iron each increased by 5%. Output of ingots in 1952 ncreased by 2½% compared with 1951. Production of rails and accessories increased by 84% in 1952. Details of other production and increases in social amenities are given.—R. S.

MISCELLANEOUS

Sheffield Laboratories of the British Iron and Steel Research Association. (*J. Iron Steel Inst.*, 1954, **176,** Feb., 167–172). [This issue].

Wide Range Metering of Oxygen in Steel Plant. M. S. Gare. (*Iron Steel Eng.*, 1953, **30,** July, 74–81). The author describes an entirely automatic oxygen metering system which produces an integrated daily reading corrected for pressure and temperature variation. It is accurate at very low as well as at high rates of flow.—M. D. J. B.

Medical and Social Services in the French Iron and Steel Industry. (*Acier Fins Spec. Franç.*, 1953, Mar., 6–8). The improvement of medical and social services in the French iron and steel industry is outlined.—V. G.

Nomenclature of Iron and Steel Products in English, French and Spanish. (*Bol. Min. e Indust.*, 1952, **31,** Nov., pp. 444–449). [In Spanish]. Names and uses of raw materials for, and products of, the iron and steel industry are given in parallel columns in English, French, and Spanish.—R. A. R.

Courses of Professional Education for Foremen in Iron and Steel Works. K.-H. Massoth. (*Stahl u. Eisen*, 1953, **73,** Sept. 10, 1231–1235). The need for technical and professional education of foremen, existing and potential, and the methods to carry it out are discussed.—J. P.

Pravda Reports on Russia's Progress in Ferrous Metallurgy. (*Pravda*, May 27, 1953: *Metal Progress*, 1953, **64,** Aug. 125–128). A short account of metallurgical developments in Russia is presented.—B. G. B.

Man, The Metallurgist. F. T. M. White. (Australian Inst. Metals: *Australasian Eng.*, 1953, June 7, 52–61). In this Presidential Address the author reviews the history, status, training, achievements, and responsibilities of the metallurgist since the dawn of civilization. Acknowledgment is made to 33 publications as the principal sources of information.

BOOK NOTICES

ASSOCIATION POUR L'ENCOURAGEMENT À LA RECHERCHE AERONAUTIQUE. '*Congrès des Matériaux Résistant à Chaud*, 1951." La. 4to, pp. [vi +] 398. Illustrated. Paris, 1953: L'Association. (Price 5000 fr.)

This book contains all the papers presented at a Congress on Heat-Resisting Materials organized in 1951 by the Association for the Encouragement of Aeronautical Research, of Paris. The texts of the 46 technical communications and of the opening and closing discourses are reproduced, but there is no mention of any discussion.

The papers are grouped under nine headings. The first section, the Introduction, contains four pages (and the

opening speech) which deal in a general way with the problems encountered in developing materials that will withstand the high temperatures to be met with nowadays in many applications, especially jet engines and gas turbines, and the succeeding sections are devoted to the physical properties of materials at high temperatures (five papers), creep (five papers), furnaces and measuring apparatus (four papers), light and ultra-light alloys (seven papers), refractory austenitic and martensitic alloys and heat-resistng steels (seven papers), ultra-refractory metals and cermets (eight papers), carbides, graphites and ceramics (four papers), and agglomerated organo-metallic materials (two papers). Abstracts of many of these papers will appear in the *Journal*.

This collection of papers is, of course, of interest primarily to designers and constructors of aircraft engines and ancillaries, but many others, such as those concerned with the manufacture of internal combustion engines, their valves and sparking plugs, of electric resistance furnaces and their heating elements, of chemically resistant refractories, of machine tools, etc., will find matter to interest them also.

A. E. C.

Brandt, D. J. O. "*The Manufacture of Iron and Steel.*" A Comprehensive Course for Operatives. (The City and Guilds Series). 8vo, pp. x + 11–384. Illustrated. London, 1953 : English Universities Press, Ltd. (Price 15*s*.)

For some years now the British Iron and Steel Federation has encouraged the instruction of operatives in the industry in the principles and broad procedure of their work. The book under review was commissioned by the Training Department of the Federation, and it has been expressly designed as a textbook for those operatives who wish to take the Iron and Steel Operatives' Examinations of the City and Guilds of London Institute. The author is a Senior Scientific Officer of B.I.S.R.A., and the book appears under the general editorship of the Director of the City and Guilds Institute. No reader can fail to be satisfied with this triple seal on the reliability and authority of the book.

In the extent of the subject matter, the title is interpreted almost to its fullest. Not only does the author deal with the production of iron and steel, but he covers, as well, the production of castings in iron and steel ; and, under the heading of The Steel Finsihing Processes, he describes rolling, forging, tube-manufacture, and the like. A valuable section is that describing the Integration, Lay-out and Location of the British Steel Industry ; this is an aspect all too likely to be neglected by a lecturer. There are fine appendices, giving short accounts of refractories, pyrometry, mechanical testing and alloy steels, and the last gives some advice on the actual operatives' Examination.

All this is presented in a clear and simple style, and following a continuous sequence, with suitable emphasis on fundamental principles. Naturally, a certain amount of filling in by the lecturer will be required, and also by the student's own observations. The many illustrations are superbly clear and interesting : not one has the reviewer seen which does not vividly fulfil its purpose ; it is doubtful whether a better collection exists in book form.

The author has expressed the hope that the book will supply a useful introduction for University and Technical College students who are starting their metallurgical studies. There is little doubt that this will be the case, and many metallurgists will probably find the book interesting. At 15*s*. this well-produced volume is fine value and all concerned with its production are to be congratulated.

A. R. Bailey

Case, Samuel L., and Kent R. Van Horn. "*Aluminium in Iron and Steel.*" (Alloys of Iron Research Monograph Series : Frank T. Sisco, Director and Editor). 8vo, pp. xii + 478. Illustrated. New York, 1953 : John Wiley and Sons, Inc. ; London : Chapman and Hall, Ltd. (Price $8·50, or 60*s*.)

The Alloys of Iron Research Monographs, publication of which was suspended during the war, have now been resuscitated under the general direction of Dr. F. T. Sisco, the first of the new series being devoted to the effect of aluminium on iron and steel.

This volume is divided into two fairly distinct parts, of which the first is concerned with the effects of the small amounts of aluminium which may be added to molten steel for the purpose of deoxidation. The second discusses the evidence regarding the effects of aluminium as a more substantial alloying element. In Section I the deoxidation of steel by aluminium is considered together with the oxide, sulphide, and nitride inclusions resulting therefrom. The influence of such additions of aluminium in the control of grain size, on notch sensitivity and on the ageing of steel are then discussed, together with other effects, such as hardenability, spheroidization, and graphitization.

Part II is concerned with the constitution of the alloys of iron and aluminium and of parts of the ternary systems. The manufacture of the alloys ; the effect of aluminium in in heat-resisting material, in permanent magnets and in nitriding steels are then considered in turn, together with the aluminium coating of steel, and, finally, a chapter on miscellaneous uses of aluminium both in steel and in cast iron. It may be mentioned that, although the title of the book is "Aluminium in Iron and Steel", only about one-twelfth of it is, in fact, devoted to cast iron.

Work done since about 1930 has been collected and reviewed in detail, but earlier work appears often to have been overlooked, such, for instance, as that of Arnold on the effect of aluminium in increasing the crystal size of ferrite (*Journal of The Iron and Steel Institute*, 1904, No. I), of Hicks on its effects on the magnetic properties, and, in particular, of Arnold and Bolsover on the influence of aluminium additions on the mode of occurrence of sulphides (*ibid*., 1914, No. I). This is perhaps the more surprising since the book by Benedicks and Löfquist, which is mentioned, does discuss this work. In this connection it may be worth pointing out that the statement on p. 86 that Lorig and Elsea found, in 1947, that "grain-boundary precipitates lower the ductility of cast steels and cause brittle fracture," though true, was, in fact, proved by Arnold over 50 years before.

The treatment, which is at times somewhat discursive and occasionally a little polemical in attitude, is also marred occasionally by statements which are at least dubious. On p. 93, for instance, we find it said that "The structural changes occurring in steel when it is heated above its A_1 transformation temperature are now fairly well explained. Above this temperature, ferrite is transformed to gamma iron, ... "!

Taken as a whole the book is undoubtedly a valuable addition to metallurgical literature. The more recent work has been most carefully scrutinized and discussed, but to the reviewer there is just the suspicion of a feeling that the general standard of this volume is not quite so high as that of its predecessors. Further books in this series which are in course of preparation are those on nickel, manganese, titanium and, jointly, niobium, zirconium, boron, tantalum and calcium. As the editors get back ino their stride, the small defects which seem to mar this particular volume will doubtless disappear.—F. C. Thompson

Enos, George M., and William E. Fontaine. "*Elements of Heat Treatment.*" 8vo, pp. [ix +] 286. Illustrated. New York, 1953 : John Wiley and Sons, Inc.; London : Chapman and Hall, Ltd. (Price 40*s*.)

This book is not really meant for metallurgists but for engineers and "personnel in commercial plants who wish to know the principles behind heat-treatment without attending a formal course."

The authors give quite a comprehensive account of the heat-treatment of steel, cast iron and non-ferrous metals. The book is reasonably complete in itself, in that the first brief chapter acts as an introduction to the nature of metals, microstructure, and alloy equilibria, although by reason of its brevity, the treatment of the last item may be misleading. Moreover, the second chapter covers the mechanical properties of metals and their assessment. Also, in another chapter there is a short discussion of shaping and forming processes, in an effort to set the heat-treatment field in proper perspective.

The strength of the book lies in the title matter, which is well illustrated by line drawings and photomicrographs. The animated illustrations do not enhance the book. The volume is certainly satisfactory for the type of reader mentioned above, although they may find it advantageous to do some wider reading on alloy equilibria. The writing is good and the book is well produced.—A. R. Bailey

AMERICAN SOCIETY FOR TESTING MATERIALS. "*Symposium on Chemical Analysis of Inorganic Solids by Means of the Mass Spectrometer.*" Presented at the Meeting of Committee E-2 on Emission Spectroscopy, Atlantic City, N.J., June 20, 1951. (Special Technical Publication No. 149.) 8vo, pp. [iv +] 35. Illustrated. Philadelphia, Pa., 1953: The Society. (Price $1.25)

BENSIMON, M. R. "*Les Matériaux Métalliques.*" Tome 1. "*Generalités. Metallographie. Methodes physiques d'études des transformations.*" La. 8vo, pp. 104. Illustrated. Paris, 1953: Editions Eyrolles. (Price 780 fr.)

BERGHEZAN, AUREL. "*Les Modifications de Structure du Cristal Métallique et leur Influence sur la Cinétique du Durcissement Structural des Solutions Solides d'Aluminium.*" Préface de G. Chaudron. (Publications Scientifiques et Techniques du Ministère de l'Air, No. 283.) La. 8vo, pp. [xi +] 95. Illustrated. Paris, 1953: Service de Documentation et d'Information Technique de l'Aéronautique. (Price 1200 fr.)

BERNAL, J. D. "*Science and Industry in the Nineteenth Century.*" 8vo, pp. xi + 230. Illustrated. London, 1953: Routledge and Kegan Paul, Ltd. (Price 12*s.* 6*d.*)

BRITISH CERAMIC SOCIETY. "*Ceramics—A Symposium.*" Arranged and Edited by A. T. Green and Gerald H. Stewart. La. 8vo, xix + 877. Illustrated. Stoke-on-Trent, 1953: The Society. (Price 45*s.* to Members, 65*s.* to Non-Members)

BRITISH STANDARDS INSTITUTION. B.S. 558 and 564: 1953. "*Nickel Anodes and Nickel Salts for Electroplating.*" 8vo, pp. 26. London, 1953: The Institution. (Price 4*s.*)

BRITISH STANDARDS INSTITUTION. B.S. 974: 1953. "*Symbols for Use on Flow Diagrams of Chemical and Petroleum Plant.*" 8vo, pp. 17. Illustrated. London, 1953: The Institution. (Price 3*s.* 6*d.*)

BRITISH STANDARDS INSTITUTION. B.S. 1121B : 1953. "*Recommended Method for the Spectrographic Analysis of Low Alloy Steels.*" 8vo, pp. 21. Illustrated. London, 1953: The Institution. (Price 5*s.*)

BRITISH STANDARDS INSTITUTION. B.S. 1224: 1953. "*Electroplated Coatings of Nickel and Chromium.*" 8vo, pp. 14. Illustrated. London, 1953: The Institution. (Price 2*s.* 6*d.*)

BRITISH STANDARDS INSTITUTION. B.S. 2035: 1953. "*Cast Iron Flanged Pipes and Flanged Fittings.*" 8vo, pp. 39. Illustrated. London, 1953: The Institution. (Price 6*s.*)

BRITISH STANDARDS INSTITUTION. S.P. 106, October, 1953. "*British Standard Specification for Swaged Cable-End Assemblies (UNF Threads) for Preformed Non-Corrodible Steel Wire Rope for Aircraft (Suitable for Use with British Standard W.11 Wire Rope).* Folio, pp. 10. Illustrated. London, 1953: The Institution. (Price 5*s.*)

CONWAY, C. G. "*Heat-Resisting Steels and Alloys.*" A Concise Data Book giving the high-temperature properties of commercial steels and alloys. 8vo, pp. vii + 160. Illustrated. London, 1953: George Newnes, Ltd. (Price 25*s.*)

"*Fifth Empire Mining and Metallurgical Congress, Australia and New Zealand*, 1953." Publications, six volumes. 8vo. Illustrated. Melbourne, 1953: Office of the Congress and Australasian Institute of Mining and Metallurgy.

Volume I. "*Geology of Australian Ore Deposits.*" A Symposium arranged by a Committee of the Australasian Institute of Mining and Metallurgy (Inc.) and edited by A. B. Edwards. Pp. xiv + 1290. (Bound volume 84*s.* Australian)

Volume II. "*Mining Methods in Australia and Adjacent Territories.*" Edited by R. Pitman Hooper and A. B. Black. Pp. xv + 374. (Price 40*s.* Australian)

Volume III. "*Ore Dressing Methods in Australia and Adjacent Territories.*" Edited by H. H. Dunkin. Pp. ix + 317. (Price 27*s.* Australian)

Volume IVA. "*Extractive Metallurgy in Australia. Ferrous Metallurgy.*" Edited by J. C. Richards. Pp. xii + 217. (Price 35*s.* 6*d.* Australian)

Volume IVB. "*Extractive Metallurgy in Australia. Non-Ferrous Metallurgy.*" Edited by Frank A. Green. Pp. ix + 272. (Price 25*s.*)

Volume V. "*Australian Mining and Metallurgy. Miscellaneous Features and Practices.*" Edited by A. W. Norrie. Pp. x + 253. (Price 25*s.*)

Volume VI. "*Coal in Australia.*" A Symposium Arranged by a Committee of the Australasian Institute of Mining and Metallurgy (Inc.) and edited by a Committee of the Sydney Branch. Pp. x + 832. (Price 60*s.* Australian)

GÖHRE, ERNST. "*Leistungssteigerung und Ausschussminderung in der Stanzerei.*" Pp. 68. Illustrated. München, 1953: Carl Hanser. (Price DM 13.50)

GRAMBERG, A. "*Technische Messungen bei Maschinenuntersuchungen und zur Betriebskontrolle.*" Siebente, neubearbeitete Auflage. Pp. x + 445. Illustrated. Berlin, Göttingen, Heidelberg, 1953: Springer-Verlag. (Price DM 36.–)

HALLIMOND, A. F. "*Manual of the Polarizing Microscope.*" 2nd edition. 8vo, pp. 204. Illustrated. York, 1953: Cooke, Troughton and Simms, Ltd. (Price 15*s.*)

HOOD, T. A. "*High-Temperature Alloys: A Bibliography.*" (Commonwealth of Australia, Defence Research Laboratories, Information Circular No. 17). Folio, pp. 105. Manibyrnong, Victoria, 1953: The Laboratories. (Mimeographed)

"*The Instrument Manual.*" Second edition. La. 8vo, pp. vii + 628. Illustrated. London, 1953: United Trade Press, Ltd.

JENKINS, I. "*Les Atmosphères Controlées dans le Traitement Thermique des Métaux.*" Traduit par J. Bernot. Préface de l'édition française de Maurice Rey. La 8vo, pp. xii + 522. Illustrated. Paris, 1953: Dunod. (Price 4800 fr.)

KITTEL, CHARLES. "*Introduction to Solid-State Physics.*" 8vo, pp. xiii + 396. Illustrated. New York, 1953: John Wiley and Sons, Inc.; London: Chapman and Hall, Ltd. (Price $7.00 or 56*s.*)

KOPPENBERG, HEINRICH, and WERNER WENZEL. "*Die Sauerstoffmetallurgie der Schachtofenprozesse.*" 8vo, pp. 237. Illustrated. Düsseldorf, 1953: Verlag Stahleisen m.b.H. (Price DM 18.–)

ORGANISATION EUROPÉENNE DE COOPÉRATION ECONOMIQUE. "*Les Procédés Nouveaux de Cokéfaction.*" La. 8vo, pp. 60. Illustrated. Paris, 1953: The Organisation.

PARRISH, W., B. W. IRWIN, and M. G. EKSTEIN. "*Data for X-Ray Analysis.*" Volume I. "*Charts for Solution of Bragg's Equation* (α versus θ *and* 2θ)," by W. Parrish and B. W. Irwin. La. 8vo, pp. [vii +] 99. Illustrated. Volume II. "*Tables for Computing the Lattice Constant of Cubic Crystals,*" by W. Parrish, M. G. Ekstein, and B. W. Irwin. La. 8vo, pp. vii + 81. Eindhoven, Holland, 1953: N. V. Philips' Gloeilampenfabrieken. (Price 15*s.* each volume)

POMEY, J. "*Les Couches Superficielles des Pièces Métalliques de Machines.*" IIIe Partie: "*Durcissement superficiel et précontrainte.*" La. 8vo, pp. 130. Illustrated. Sèvres, Paris, 1953: Editions de la "Revue d'Optique. (Mimeographed. Price 1000 fr.)

PRITULA, V. A. "*Cathodic Protection of Pipelines and Storage Tanks.*" Translated from Russian. (Department of Scientific and Industrial Research.) 8vo, pp. vi + 160. Illustrated. London, 1953: H.M. Stationery Office. (Price 10*s.*)

STOUT, ROBERT D., and W. D'ORVILLE DOTY. "*Weldability of Steels.*" 8vo, pp. 381. New York, 1953: Welding Research Council. (Price $6.50)

VAN SANTEN, G. W. "*Introduction to a Study of Mechanical Vibration.*" With a Foreword by B. E. Noltingk. Translated into English by G. Ducloux. 8vo, pp. xvi + 296. Illustrated. Eindhoven (Holland), 1953: N. V. Philips' Gloeilampenfabrieken.